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Vol. 28 : No. 194 NOVEMBER, 1961 Price 2/6

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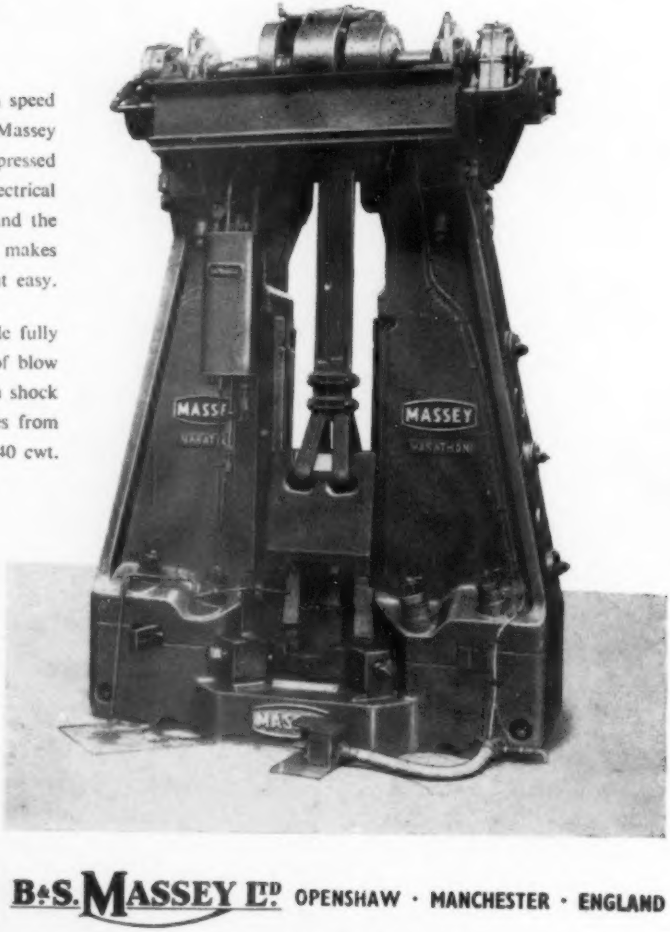

20 *cwt. Marathon Drop Hammer*

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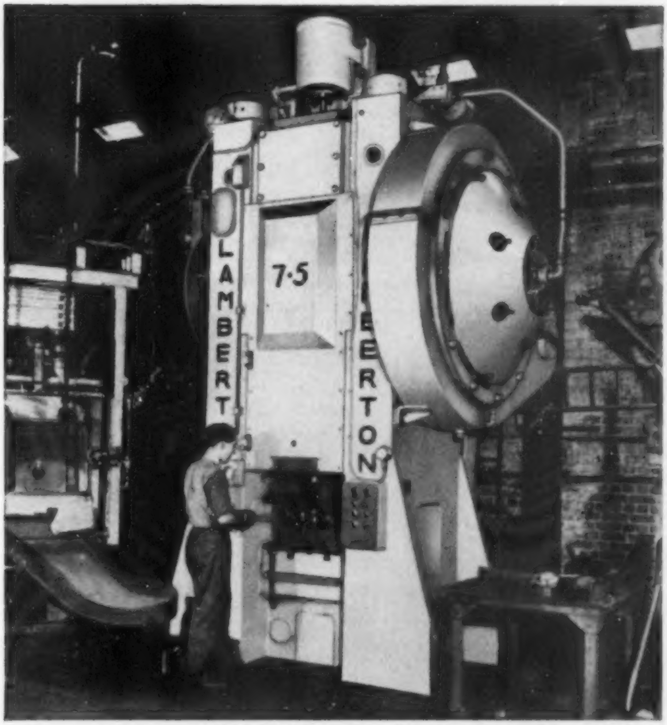

Photograph by courtesy of The Austin Motor Company Limited

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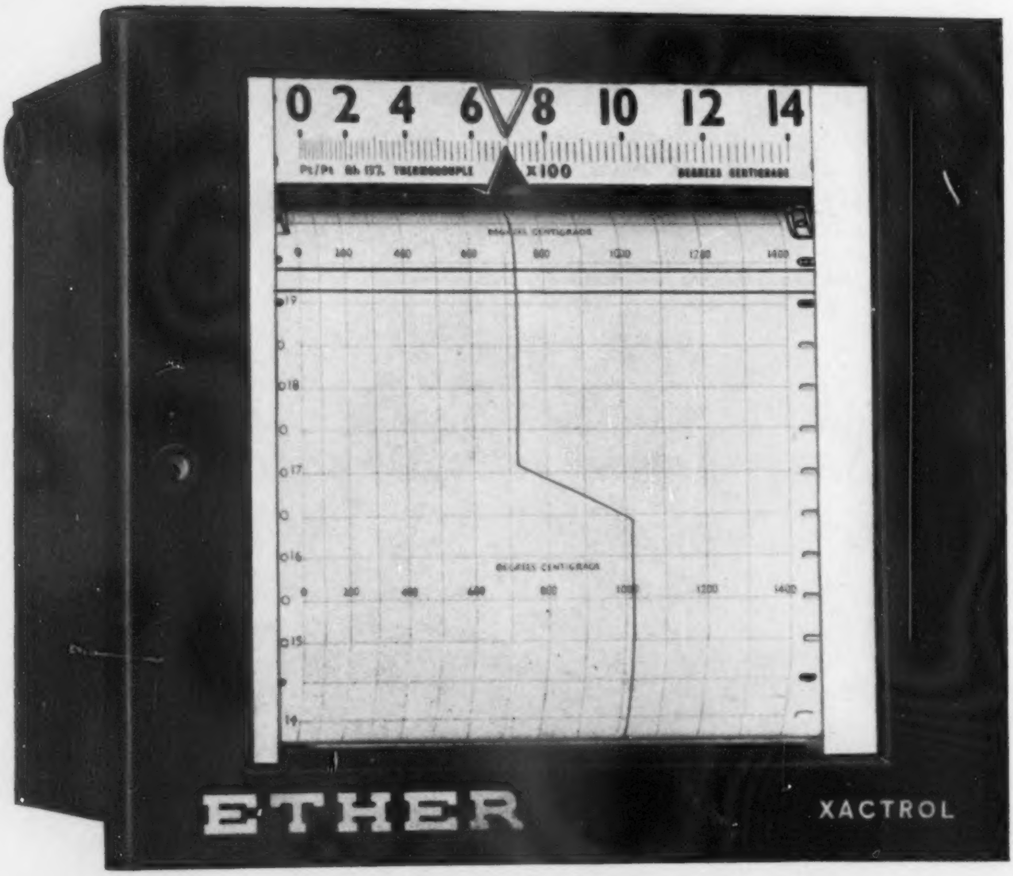

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PATENTS APPLIED FOR

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Single-point Potentiometer Recording Controllers

6″ calibrated scale * Standard ranges: 2mV-100mV span
Pen speed: 1, 2 or 4 secs. across chart * Chart life: 1 month at 1 inch per hour.

Available with (i) Two-position on/off control (ii) Anticipatory control (iii) Proportioning (electrical) control (iv) Three-position control (v) Programme control (vi) Proportioning (motorised) control

SERIES 2050
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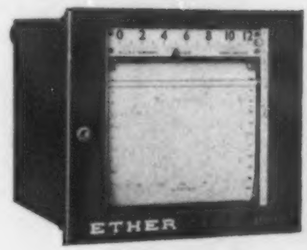
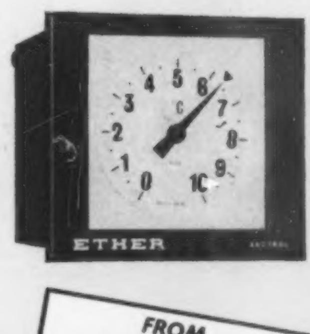


6″ calibrated scale * Balancing motor totally de-energised throughout printing and indexing cycles *
Standard ranges: 5mV-100mV span *
Pen speed: 2 secs. across chart * Printing speed: 6 secs. per point

Available with (i) Two-point record (ii) Three-point record (iii) Six-point record.

SERIES 2500
Single-point Potentiometer Indicating Controllers



18″ calibrated scale * Standard ranges: 5mV-100mV span * Response speed: 2 secs. across chart * Bold easy-to-read scale and pointer.

Available with (i) Two-position on/off control (ii) Anticipatory control (iii) Proportioning (electrical) control (iv) Three-position control (v) Programme control (vi) Proportioning (motorised) control

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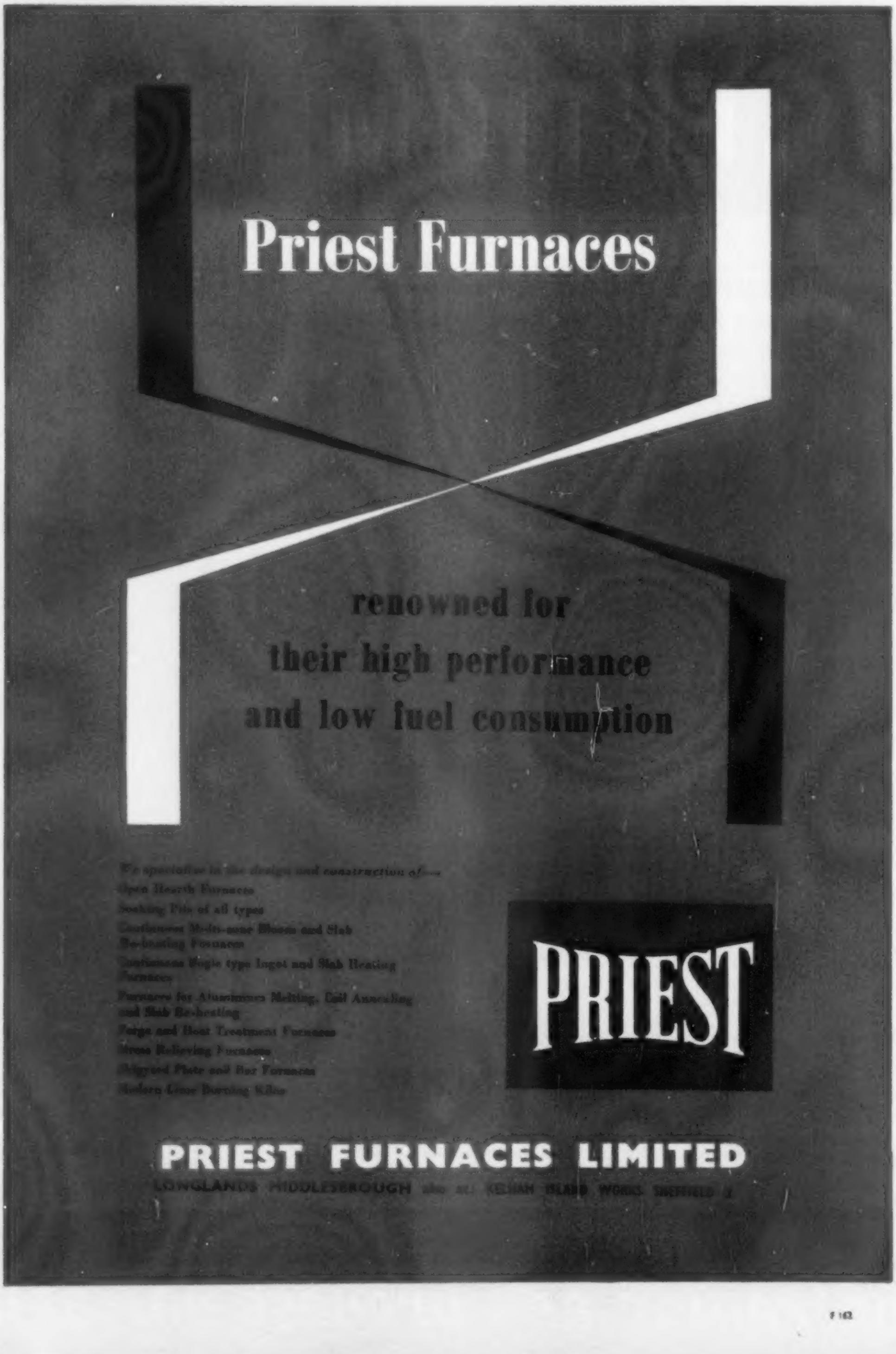
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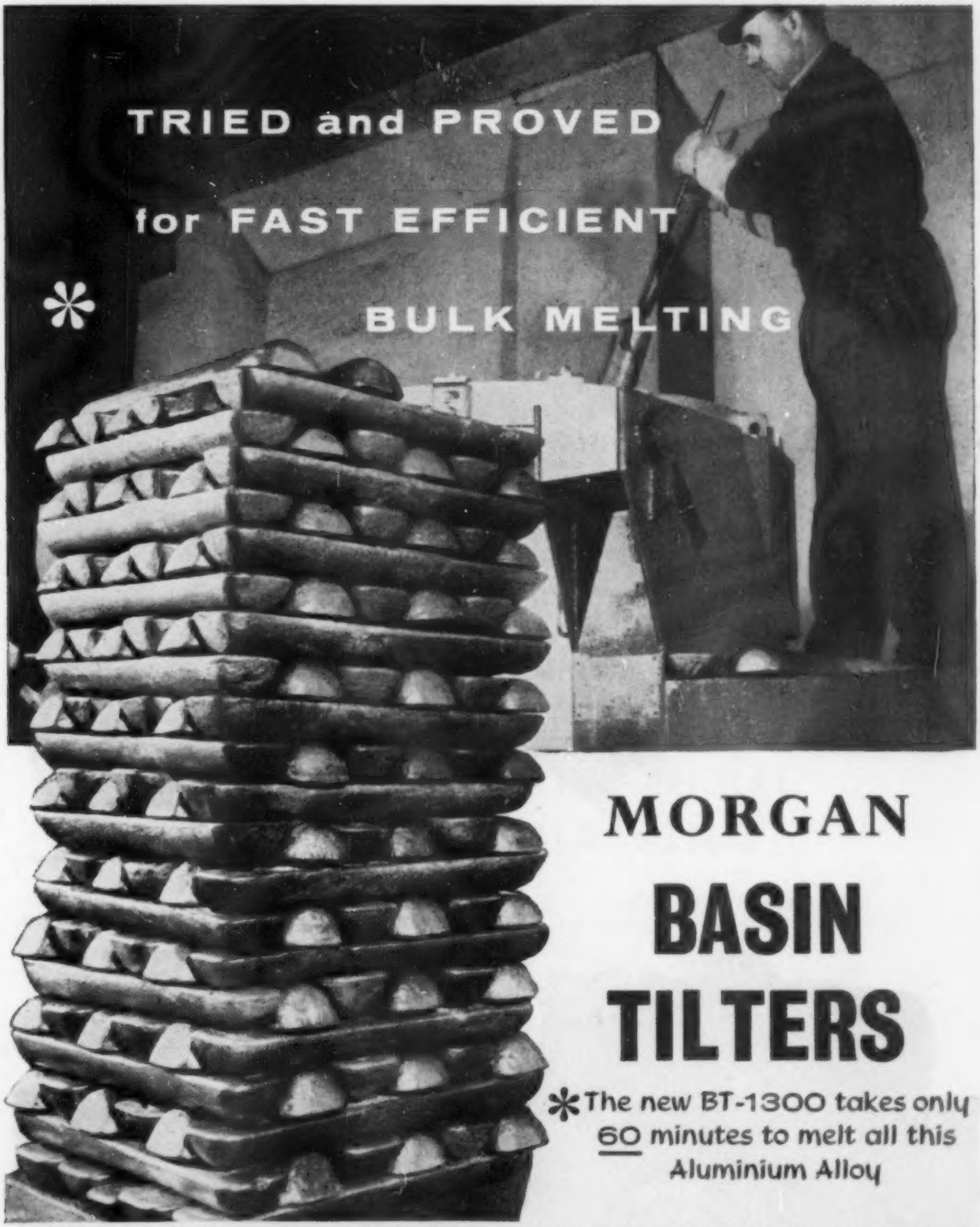

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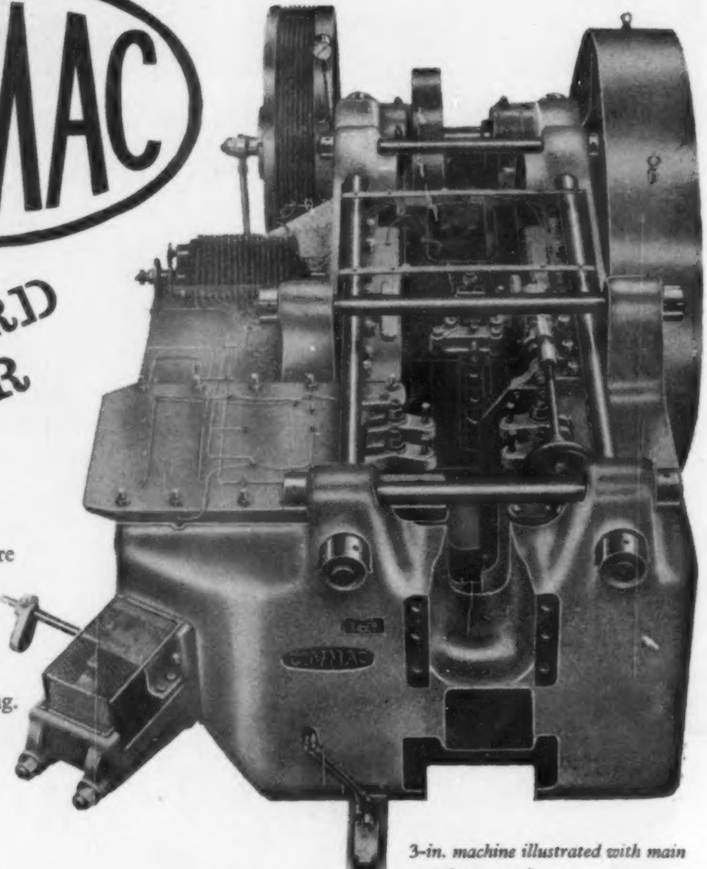

3-in. machine illustrated with main guards removed

Covmac also make Automatic Forging Machines, as well as machines for making balls up to and including 4-in. diameter. In addition there are three sizes of Hot Milling and Sawing Machines.

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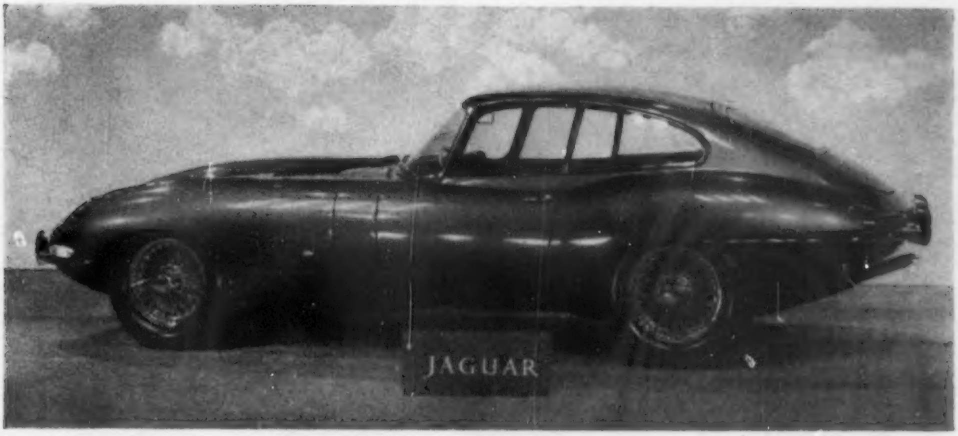

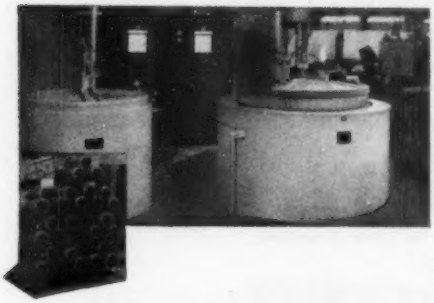

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Jaguar Model E

combines the superb performance of a racing car with the comfort and reliability of a family saloon.

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WE are indebted to the Jaguar Company for the above photographs of the new Model E and part of their Homocarb furnace installation. Homocarb with its twin advantages of forced circulation and exact temperature/time control has won world-wide recognition as will be seen from the following list of some important users.

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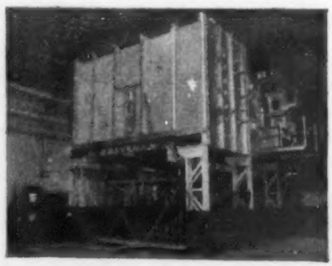

This oven furnace

is just one type of heat recirculation furnace available. Others include Bogie Hearth, Vertical & Horizontal Pit types. A standard range of vertical air recirculation furnaces is also available for tool room use. Temperature range 100-750°C.

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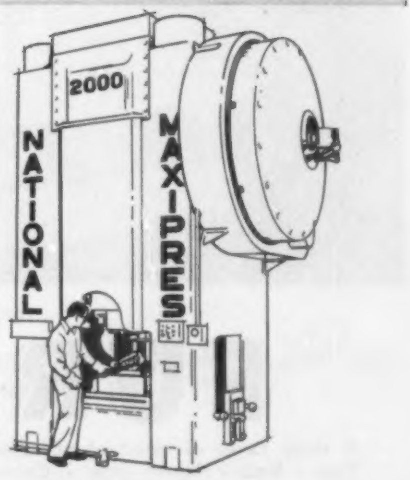

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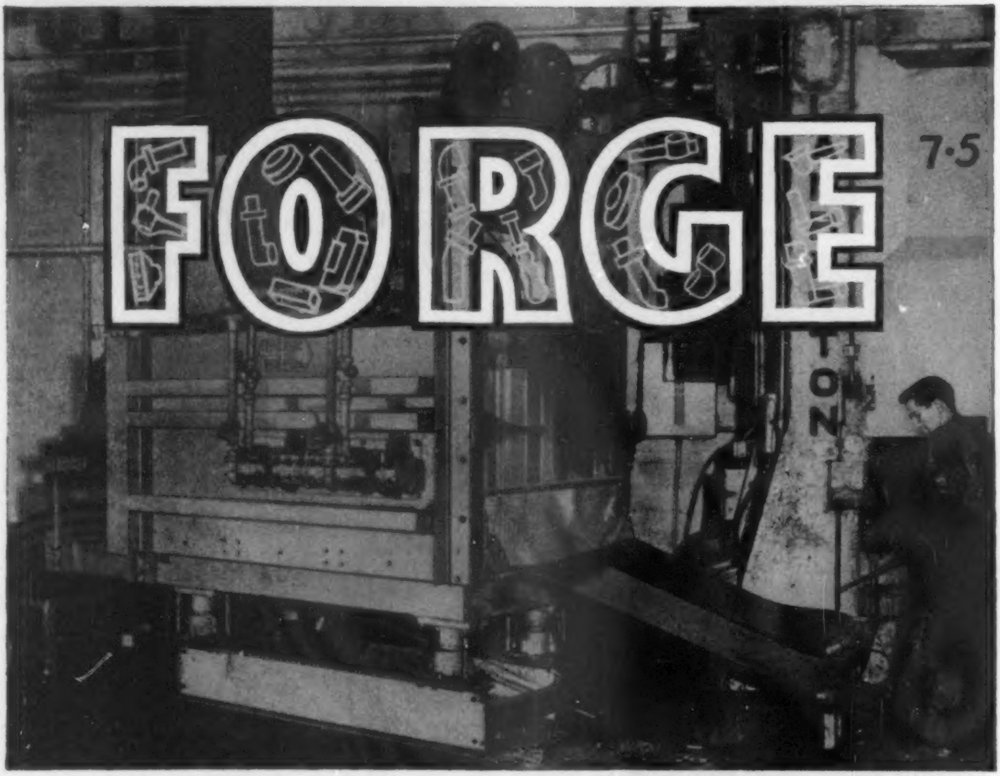

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A wide range of standard or tailor-made mechanised forge furnaces is available. Pusher Type - Rotary Hearth Type - Conveyorised Bar End Heating, with GGC scale-free heating system or fired by gas or oil. The photograph above illustrates a small Thermic Magazine Feed Pusher furnace fired by CC Burners. Output: 7-cwts. of small billets per hour.

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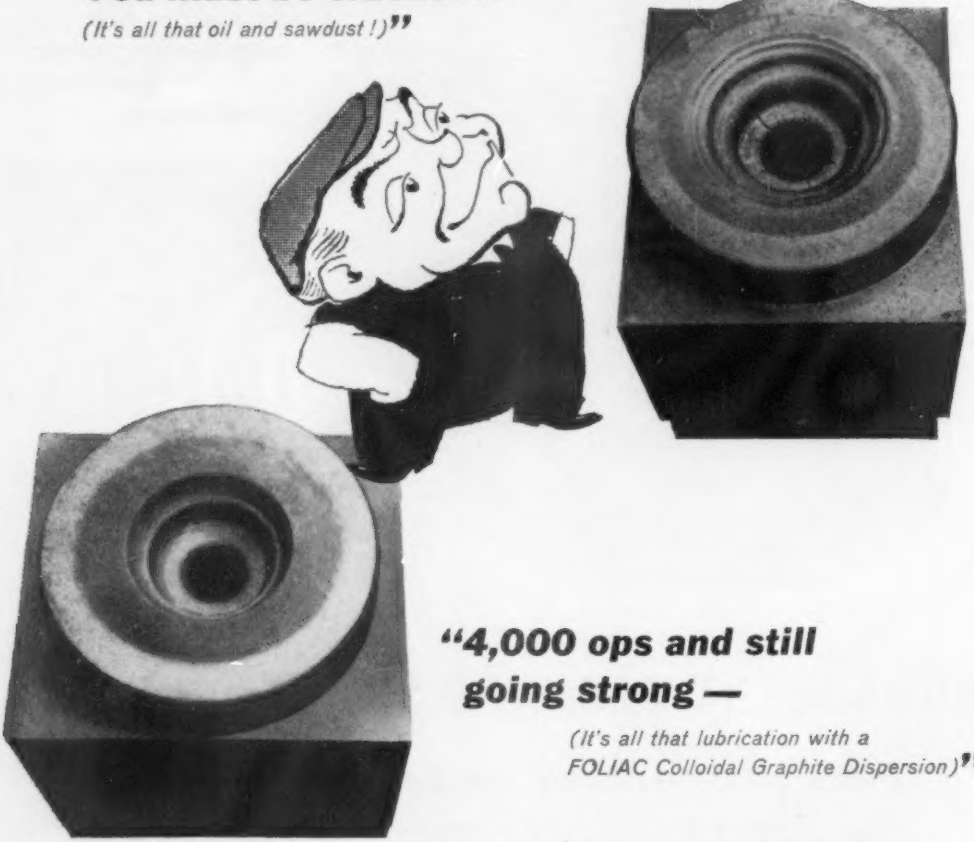
You must be cracked...**
(It's all that oil and sawdust!)"



"4,000 ops and still going strong —
(It's all that lubrication with a FOLIAC Colloidal Graphite Dispersion)"

Foliac Colloidal Graphite supersedes the traditional methods of die lubrication. It not only eliminates metal-to-metal contact, but protects the surfaces of the dies from corrosion, cracking and burning. It provides a tough, self-lubricating film over the entire die surface that reduces friction and improves metal flow. The result is a consistently well finished product from a die that lasts twice as long—or longer. Why not ask our representative to call to discuss specific problems, and advise on applications.

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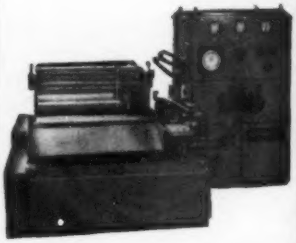
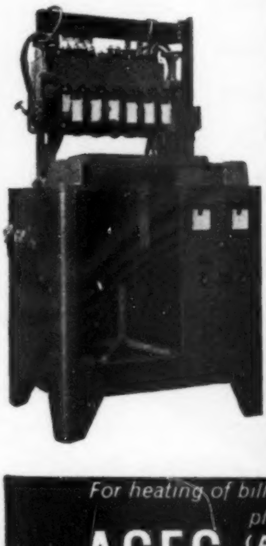
ACEC FORGES AHEAD . .

160 kw 8 k/c "Transaxe" type heater for heating the ends of pins.



160 kw 8 k/c Screw spike heater



Induction Heating applied to large scale production in modern forges has many advantages over traditional heating methods, and is being widely adopted by those with an eye to the future.

From the production angle INDUCTION HEATING reduces the heating time, the formation of scale, fire losses and rejects. Regular production rates can be maintained, giving increased tool life and automatic feeding is made possible.

Labour costs are greatly reduced because kindling and cleaning operations are rendered unnecessary. Fuel handling is eliminated, automation becomes a practical possibility and hygienic working conditions a reality.

ACEC INDUCTION HEATERS are compact and can be installed exactly where needed. Accurate temperature control puts the final touch to the guarantee of a high quality product.

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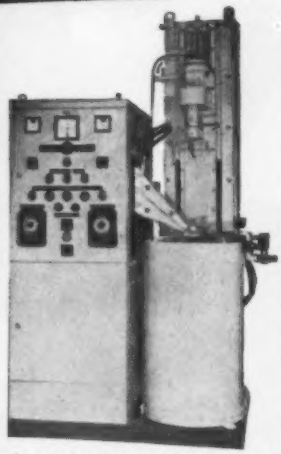

Above: Automatic machine for surface hardening of small cylindrical parts

Below: 100 kw 4 k/c heater for tubes

For heating of billets, pins, tubes and bars of all shapes and sizes, please get in touch with

ACEC (Ateliers de Constructions Electriques de Charleroi)
Cromwell House, Fulwood Place, London, W.C.1

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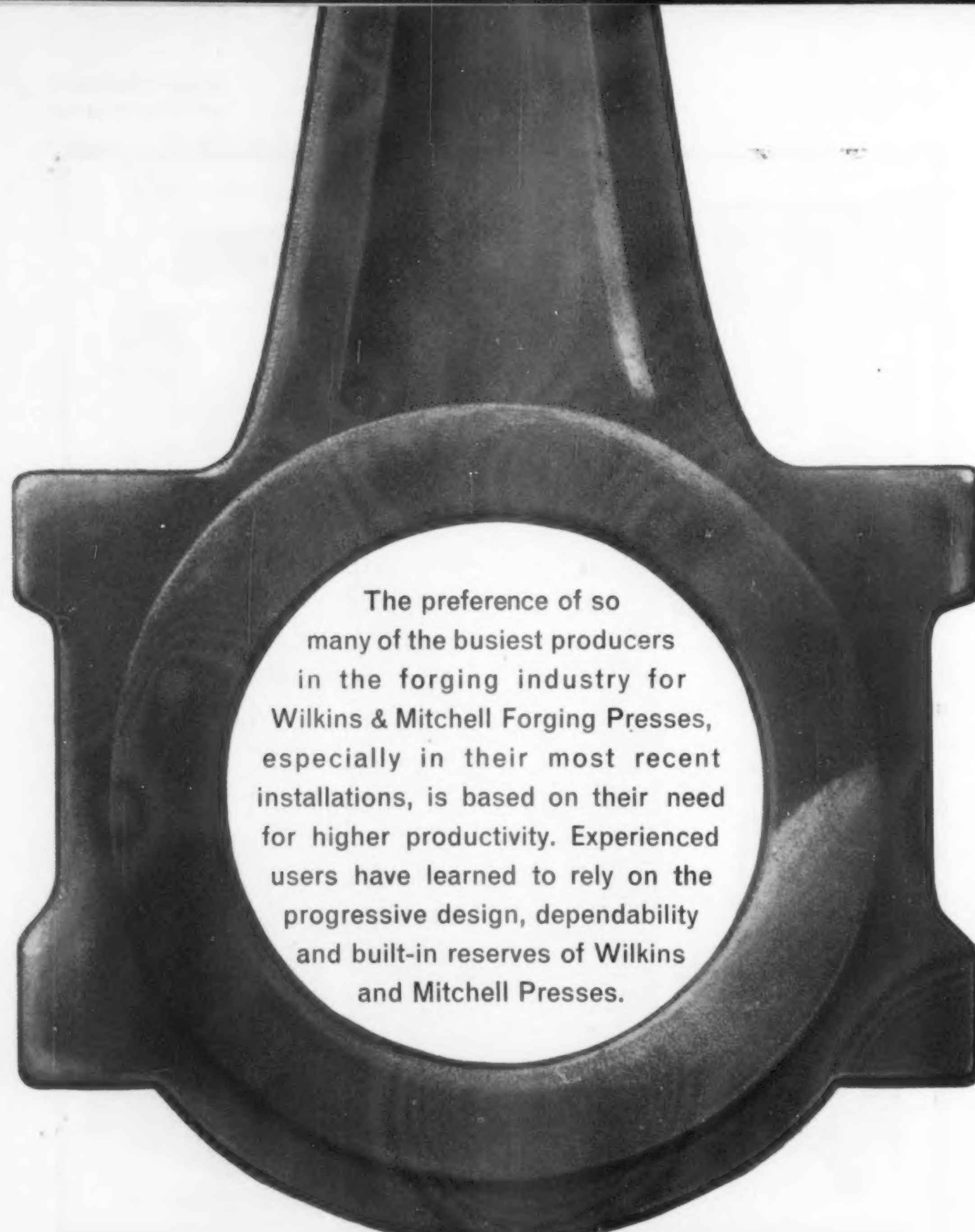

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STRIPPING PRESSES
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BILLET SHEARS
of advanced design
and related efficiency

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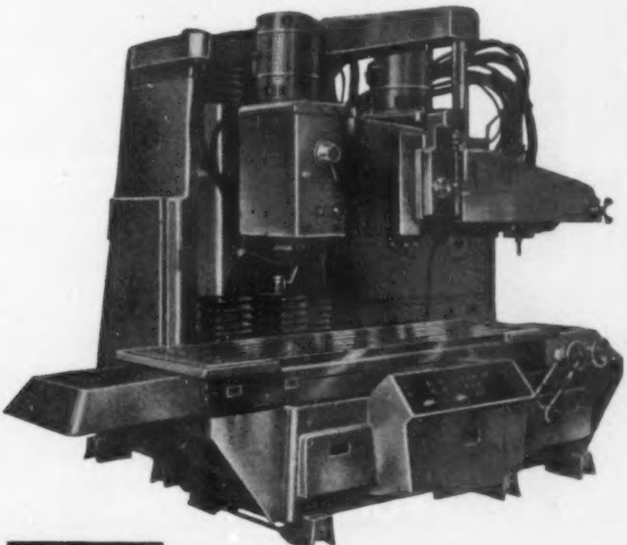

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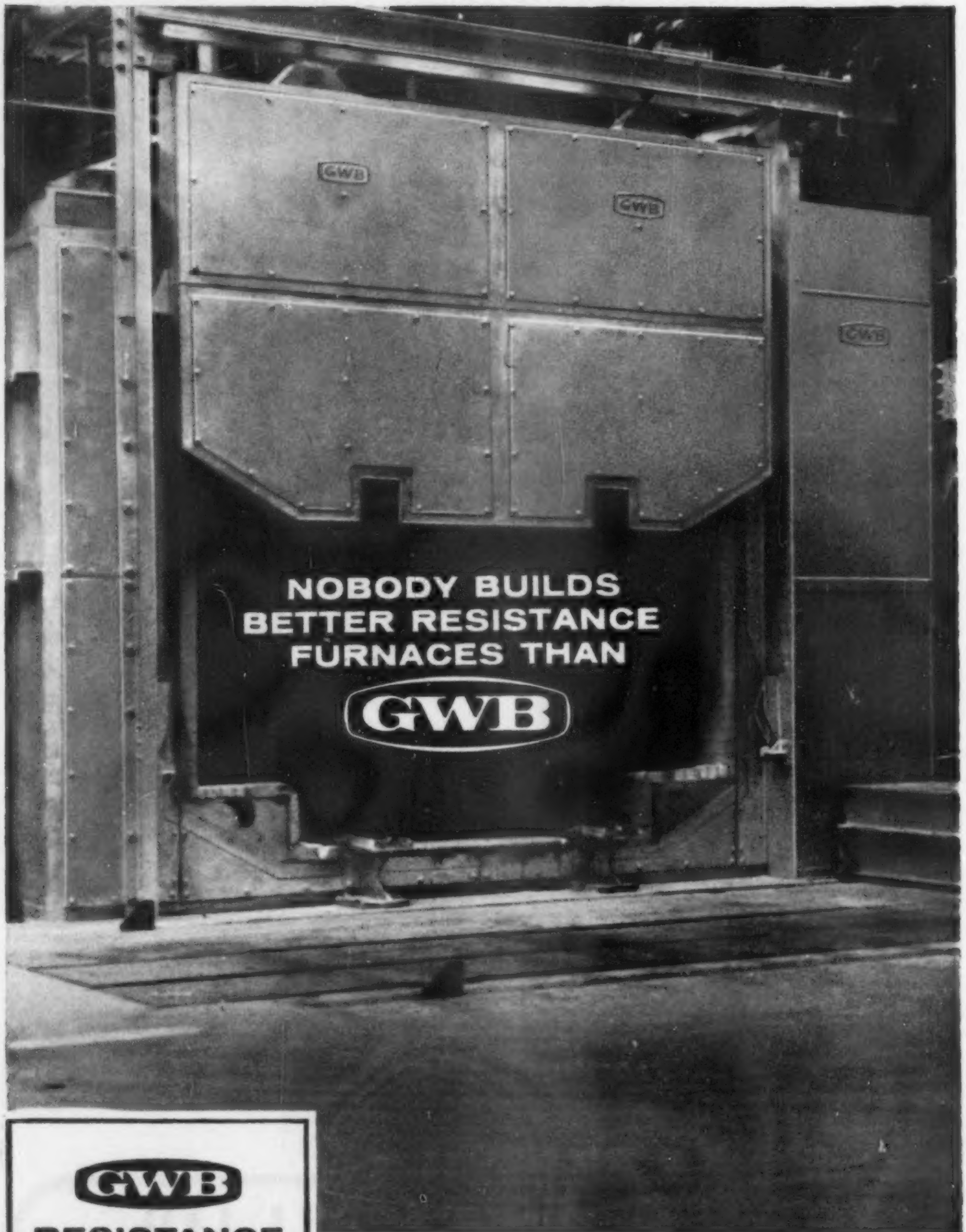

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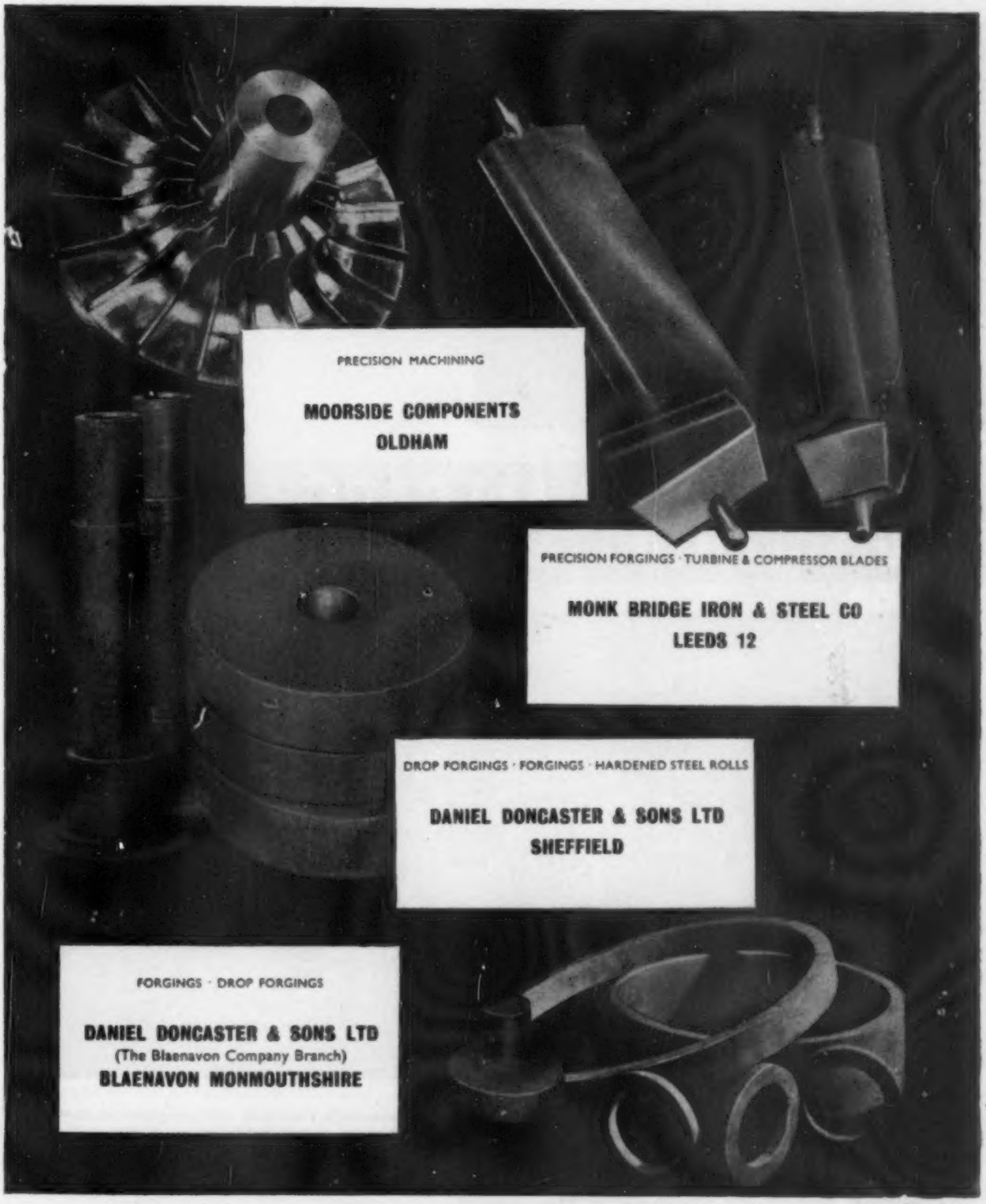

The Daniel Doncaster companies are integrated to bring together modern resources and accumulated skills so that we can carry work through from the forging stage to the highest standards of finished machining.

We are equally ready to take up the work at any stage — to let you have the rough forging or to precision machine your own components.

We are specially interested in manipulating difficult materials — complex alloy steels, titanium, Nimonic, heat resisting materials.

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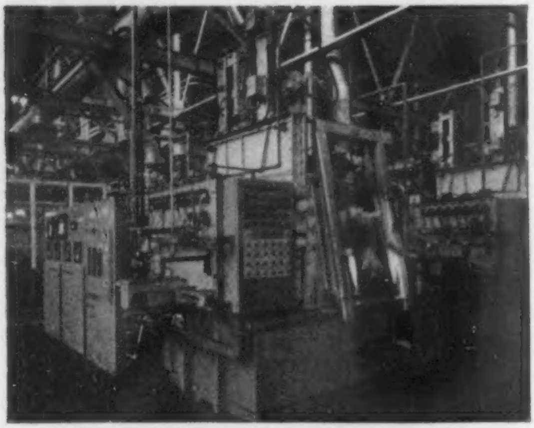
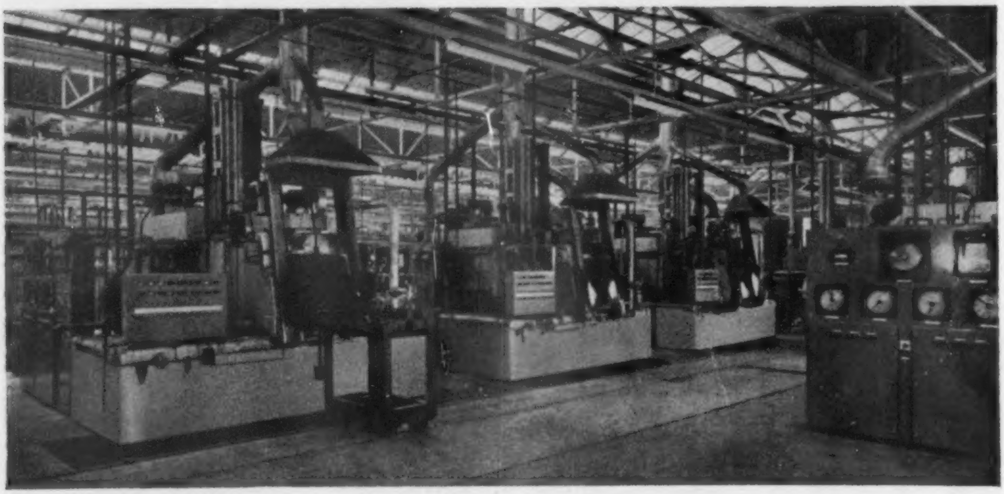


Two "Allcase" Furnaces at the works of a leading motor manufacturer, used for carburizing and carbonitriding steering and other parts, requiring various case compositions and depths. The hearth area of each furnace is 3′ 0″ x 2′ 0″ with 1′ 6″ permissible depth of charge, and is designed to accommodate gross charge weights varying from 900 lb. at 750°C. to 500 lb. at 950°C.

the versatile "ALLCASE" furnace



The "Allcase" Furnaces illustrated form part of a battery of fully-automatic sequence and programme controlled Furnaces used for Gas Carburising, Carbonitriding, and Reheating of various automobile components.

PROCESSES: With an operating temperature range of 1,400 to 1,750°F. the following controlled atmosphere processes can be carried out.

1—Gas Carburizing. 2—Hot Oil Marquenching. 3—Clean Hardening. 4—Dry Cyaniding or Carbonitriding. 5—Carbon Restoration. 6—Homogeneous Carburizing. 7—Clean Annealing.

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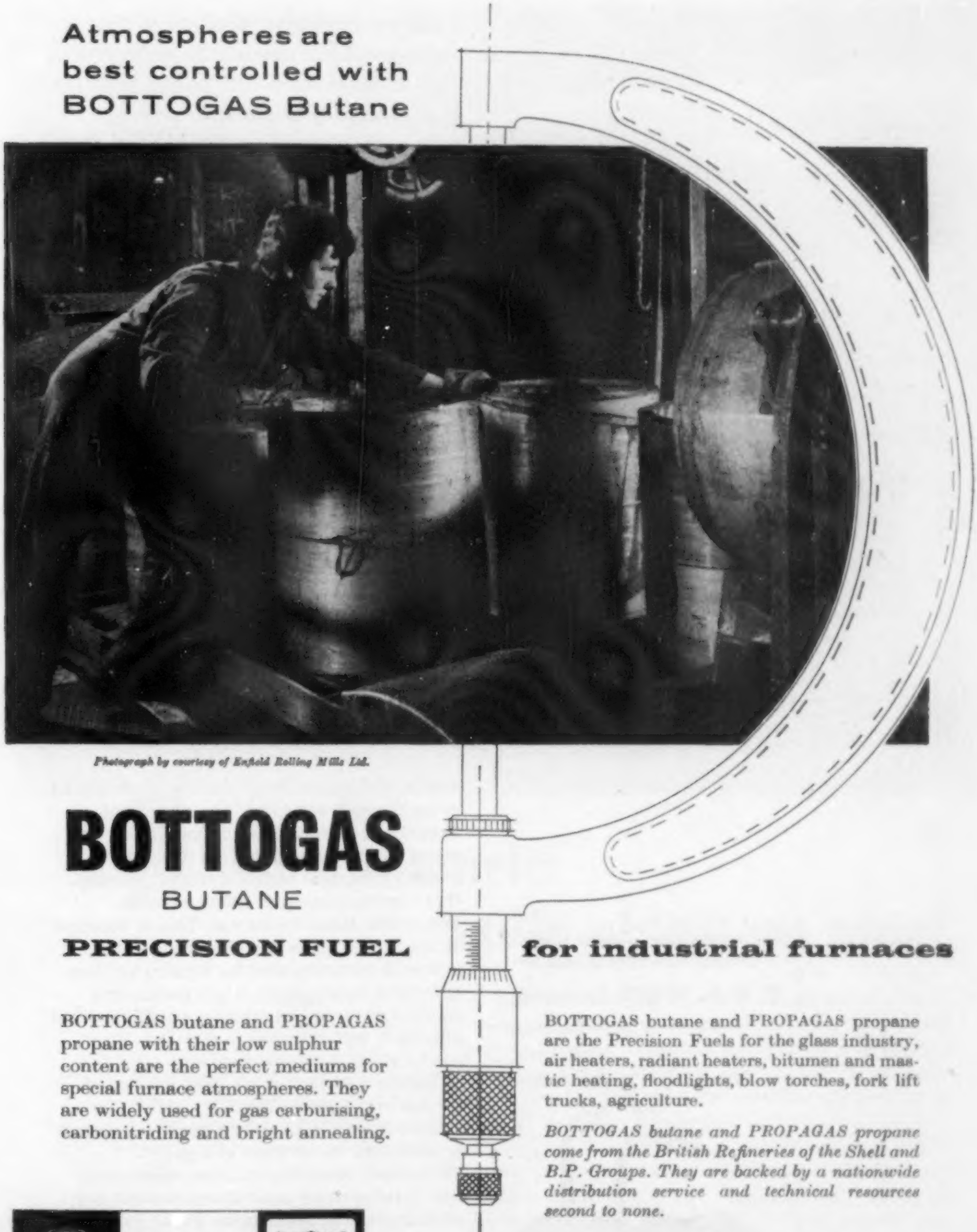
ASSOCIATED WITH SURFACE COMBUSTION CORPORATION, TOLEDO, U.S.A.

Atmospheres are best controlled with BOTTOGAS Butane



Photograph by courtesy of Enfield Rolling Mills Ltd.

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PRECISION FUEL **for industrial furnaces**

BOTTOGAS butane and PROPAGAS propane with their low sulphur content are the perfect mediums for special furnace atmospheres. They are widely used for gas carburising, carbonitriding and bright annealing.

BOTTOGAS butane and PROPAGAS propane are the Precision Fuels for the glass industry, air heaters, radiant heaters, bitumen and mastic heating, floodlights, blow torches, fork lift trucks, agriculture.

BOTTOGAS butane and PROPAGAS propane come from the British Refineries of the Shell and B.P. Groups. They are backed by a nationwide distribution service and technical resources second to none.

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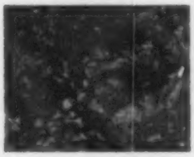
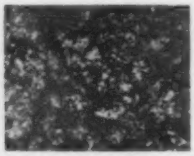
If you are involved in all or any of these processes then it may well be worth your while to get in touch with E. & A. West Ltd. In all probability they may be able to solve some of your problems in these fields or suggest more economical or efficient methods. Their Technical Advisory Service is at the disposal of Industry and will be glad to get to grips with your particular problem. The answer may even be in their impressive library of "case histories". Please do not hesitate to call them in. This service is freely given and non-obligatory.

× 150



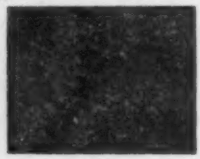
18/8 Stainless Steel surface after two hours in Nitric Hydrofluoric Acid bath. Over pickling has resulted in etched crystalline surface.



18/8 Stainless Steel surface after two hours in Ferric Sulphate Hydrofluoric Acid bath. Over pickling has had no effect on the metal surface.

× 40

Patches of red scale left on brass after pickling in 7% Sulphuric Acid.



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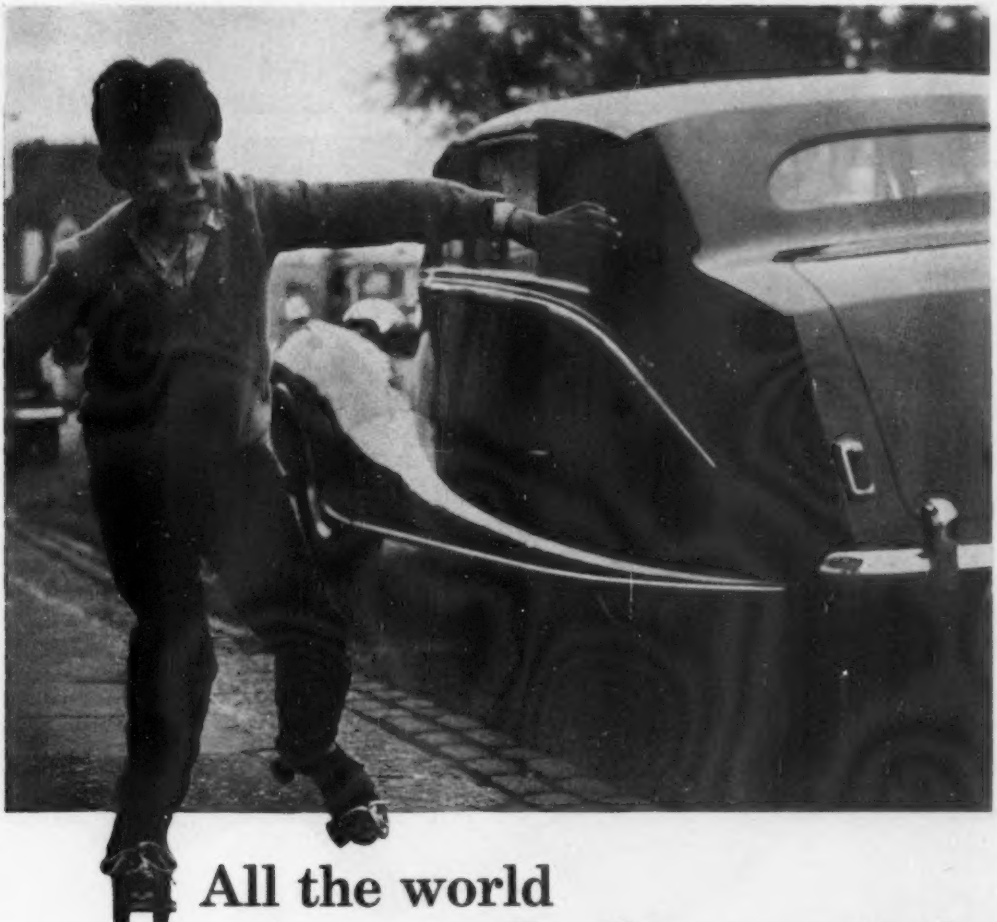

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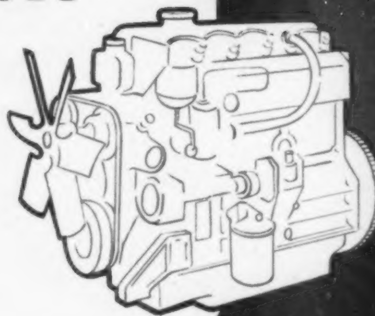

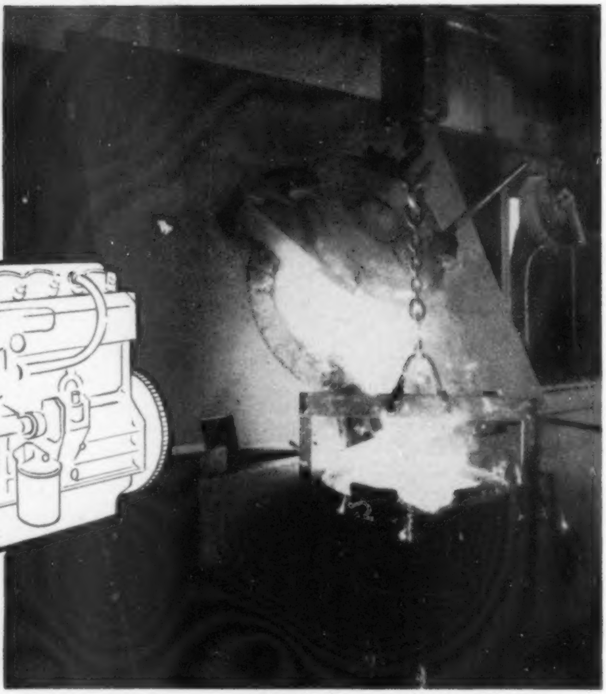

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SM/BE 6506

November, 1961

Vol. 28, No. 194

Metal treatment

and Drop Forging

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Official organ of the
NATIONAL ASSOCIATION OF DROP FORGERS AND STAMPERS
President S. JOHNSON
Director A. L. STUART TODD, C.B.E., J.P.
Grove Hill House, 245 Grove Lane
Handsworth, Birmingham 20
Telephone NORthern 3311-3

Published on the 15th of each month by
INDUSTRIAL NEWSPAPERS (FUEL AND METALLURGICAL) LIMITED
AN INDUSTRIAL NEWSPAPERS PUBLICATION
John Adam House
17-19 John Adam Street
Adelphi, London, W.C.2
Telephone TRAfalgar 6171 (10 lines)
Telegrams Zacatecas, Rand, London
Subscription Terms Home and Overseas 30*s.* per annum prepayable
Single copies 2*s.* 6*d.* (3*s.* post free)

CONTENTS

This journal is devoted to metals—ferrous and non-ferrous—their manufacture, properties, heat treatment, manipulation, testing and protection, with research work and development in all these fields

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W.B. 101

A science degree

WHEN the young graduate in classics or history has to consider his future career, he knows very well that there will be few chances of his finding employment in his own field of study. The science graduate, however, is in quite a different situation. The demand for qualified men in almost all the scientific professions still exceeds the supply and so the choice of career offers no problem. Indeed, it might almost be said that increasingly narrow specialization at university or college of technology is tending to eliminate choice altogether, and the young specialist trying to enter another field might well be looked at askance.

The shortage of scientific manpower has been with us for so many years now that the idea of a possible future surplus is very hard to envisage. Yet this is the conclusion reached by the Committee on Scientific Manpower in a statistical assessment of demand and supply for the next 10 years. ('The long-term demand for scientific manpower.' H.M.S.O. 1961). According to this report, by 1965 the supply and demand of scientific manpower should be in balance, and thereafter a surplus may exist although this may be balanced by a slight shortage of technologists.

In the view of the Committee, the possibility of a surplus of scientists over immediate demands should be welcomed. It makes possible a rational use of the scientific disciplines and means that at long last Britain would have a supply of qualified manpower with a scientific training for management, administration and the professions generally. The Committee does not doubt that scientific education will adapt itself to this new prospect, and that, just as only a proportion of those trained in the classics have expected to find employment in their own fields, an increasing proportion of those trained in specialized scientific disciplines will obtain employment outside them.

The Committee considers that both science and the nation will benefit from this adjustment and it believes that the universities are already thinking along these lines. If the Committee's confidence is to be justified, it is essential that the universities should be ready to act vigorously before the expected surplus arises. Modifications in the training of chemists and physicists such as have already been suggested by the universities or by professional institutions could go far to correct the narrowness and rigidity of training which limit the employability of a scientist or technologist trained in a particular discipline. Colleges of technology, perhaps even more than the universities, will have a difficult task to balance the demands of industry for training in particular techniques, with the need for a broad general scientific background.

Assuming the surplus, and assuming the reorientation in education, will, in fact, the employer be willing to try the experiment of using science graduates in other occupations? It is not solely the reluctance of scientists and technologists to take up administrative appointments that has led to their scarcity in these posts. It is as much, and possibly even more, due to the unwillingness of firms to give such opportunities to a scientist or technologist. The persistent belief that a person trained in a scientific discipline is inherently unfit for management responsibilities, is a prejudice that dies hard. Educational improvements, management training courses and so on are necessary, but it is even more necessary to educate public opinion in order to overcome this prejudice. Only in this way will the full advantage of the scientific viewpoint in the commercial and industrial world be realized.

NADFS annual banquet 1961

Personalities seen at the NADFS annual dinner held at Birmingham on November 2 (names read left to right)

1 *Rt. Hon. Lord Mills,* K.B.E., *Minister without Portfolio, and the president, Mr. S. Johnson*

2 *Mr. L. R. Clements, Mr. J. Mitchell,* J.P. *(chairman, the Midland Forgemasters Association), and Mr. H. M. H. Fox*

3 *Mr. J. H. Swain, past president, Mr. D. W. Turner,* J.P. *(member of the Executive Committee of the British Iron & Steel Federation), and Mr. R. Bennett, past president*

4 *The president, Mr. A. F. Kelley (Rolls-Royce Limited), Mr. R. Steel, and Mr. G. W. Richards*

Effect of structural changes on the mechanical properties of hardenable, creep-resistant 35Ni-15Cr-3W-Ti, Al steel

LUBOMÍR ČIŽEK, DR. JAROSLAV JEŽEK and JOSEF VOBORÍL

Structural and mechanical tests on hardenable creep-resistant 35Ni-15Cr-3W-Ti, Al steel were carried out by the authors of the Research Institute for Materials and Technology, Prague, and reported in Hutnické Listy, 1961, (9)

REFERENCE has already been made in an earlier paper to the properties of creep-resistant steel Poldi AKRN.[1] It was then established that the excellent resistance of this steel to creep and relaxation, in addition to its other properties, makes possible its relatively wide application for structural components operating at temperatures around 650–675°C., and for components subject to less severe loads even up to a temperature of 700°C. At the present time, its producion has been developed to the extent that it has been possible to proceed to its use for steam turbine blades. The exceptionally good relaxation resistance ranks this steel among the best materials for bolts.

On these grounds, in our earlier papers[2, 3] we have devoted attention to the structural changes which take place in this steel during heat treatment and under load in service, with the aim of determining the relationship between the structure and the properties of this steel and of other similar materials.

Steel composition

The basis of steel AKRN is formed by iron together with 35% Ni, 15% Cr and 3% W. Precipitation hardening is made possible by the addition of Ti in a quantity of about 1·5%, and some part is also played by aluminium, which passes into the steel during steelmaking in a small quantity of up to about 0·5%.

No equilibrium diagrams for this type of steel, which might be able to give some idea of the structure, are available at present. The structural proportions can only be judged from the simpler Ni-Cr-Ti, Ni-Cr-Al and Ni-Ti-Al ternary diagrams, and from the pseudo-ternary Ni-Cr-Ti-Al diagram, compiled by Taylor and Floyd[4–7] for Nimonic alloys. The steel under investigation differs from these alloys through the fact that part of the chromium and the greater part of the nickel are replaced by iron, and there is a smaller quantity of tungsten. As Taylor[8] showed, partial replacement of the chromium by iron has very little influence on the structure, and the Ni-Cr-Fe-Al-Ti system differs only little from that previously mentioned. Likewise partial replacement of the nickel by iron is not manifested by any qualitative difference in the relative diagrams,[9] and finally Kornilov and his collaborators[10] showed that substantial changes also cannot be expected from the addition of a small quantity of W. As a comment on the results of the authors mentioned, however, it should be observed that the conclusions following from them are valid only from the aspect of the existence of corresponding equilibrium phases within a whole complex of similar alloys. In contemporary papers,[11] it is shown that the exact course of the boundaries of the phase regions differs for the individual alloys, and in this regard it is impossible to apply without discrimination the approximation stated above.

For our purpose of a preliminary assessment of the possible phases, however, on the basis of the papers mentioned, we can expect that the structural proportions of AKRN steel will be similar to those of Ni-Cr-Ti-Al alloys of the Nimonic type. This implies that, in addition to the basic γ solid solution, the presence is also possible of the γ'-phase with a basic composition Ni_3Al, having a face-centred, cubic lattice, the parameters of which differ only slightly from the parameters of the γ solid solution. This phase can dissolve titanium, so that its formula can be written as $Ni_3(Al, Ti)$. Here up to three-fifths of the atoms of aluminium can be

replaced by atoms of titanium. By this substitution of aluminium atoms by titanium atoms the difference between the parameters of the γ and γ' lattices is increased, which is favourably manifested after hardening in the resistance of the alloy to creep. If the ratio of Ti to Al is increased above the solubility limit of titanium in the Ni_3Al phase, the hexagonal η-phase starts to appear in the structure, the stoichiometric composition of which, Ni_3Ti, remains constant. The phase mentioned brings about hardening of alloys of the Nimonic type. According to papers by Soviet authors,[12] in addition further elements dissolve in the Ni_3Al phase, so that its formula may be written as $(Ni, Cr)_3$ (Al, Ti, Mo, W). In our steel a small quantity of iron can also dissolve in this phase.[13]

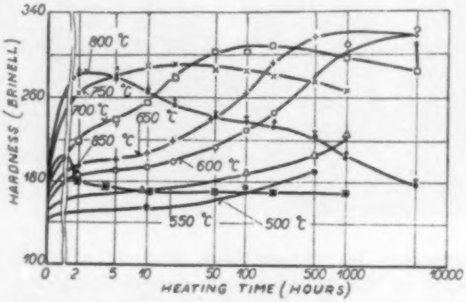
Apart from the hardening phases, γ' and η, in the structure carbides may be expected, namely in particular TiC and Cr_7C_3 or $Cr_{23}C_6$, or complex M_7C_3 and $M_{23}C_6$ carbides. In addition, in similar alloys further phases have been found, such as intermetallic Laves phases of the type A_2B, the G phase,[14] the N phase of the type Al_nA_2B,[15] and finally the σ-phase, the deposition of which can occur in the vicinity of the precipitates as a consequence of local impoverishment of the background solution in nickel. These phases do not take part in the hardening of the alloys investigated, since they precipitate predominantly at the grain boundaries, but precisely this circumstance can have a distinct influence on the resistance of these alloys to deformation at high temperatures.[16]

Effect of precipitation hardening on mechanical properties

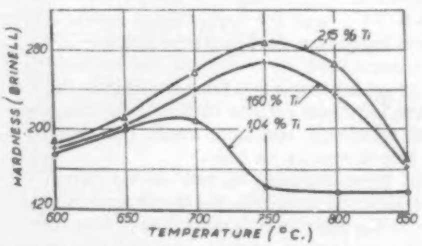
The first stage in the heat treatment of hardenable alloys is solution heat treatment, the purpose of which is the transfer of the hardening phases into the solid solution and the removal of work hardening after preceding mechanical working. During solution heat treatment grain growth also occurs, which is the limiting factor for the choice of the temperature and time of the solution heat treatment.

During research into the most suitable solution heat treatment for AKRN steel, it was established that work hardening is removed by heating at temperatures above 1,050°C., while, at the same time, the intermetallic phases are also dissolved after two hours' heating above this temperature. By creep tensile tests, it was shown that the maximum times until fracture are reached by specimens which have been subjected to two hours' heating at a temperature of 1,150°C.

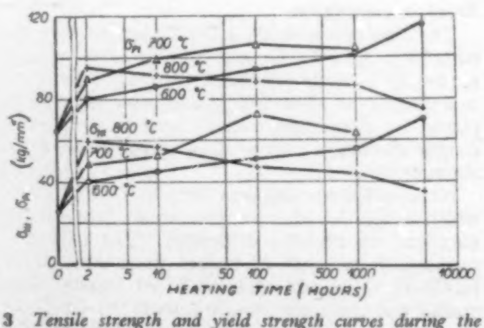
After quenching from the solution treatment temperature, we obtain a supersaturated gamma solid solution in which are located TiC carbides which have not been dissolved during heating. The steel has low hardness, tensile strength and yield strength values, but high elongation, reduction in area and notch impact strength values. For a melt, for instance, which contained 2·5% Ti, after solution heat treatment (1,150°C., 2 h., water) we obtained the following values: H_B = 140, u.t.s. 62 kg./mm.², y.p. 25 kg./mm.², δ = 45%, ψ = 65%, notch impact strength = 22 kg.m./cm.²



1 *Hardening curves of the type* AKRN *alloy containing 2·15% Ti*

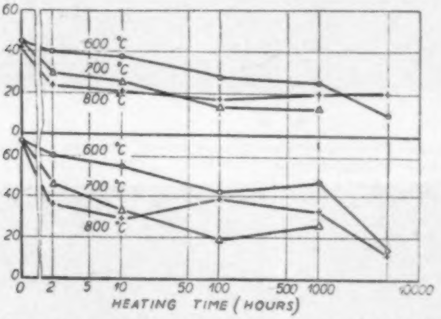


2 *Hardening curves of type* AKRN *alloys having various Ti contents (constant time of* 10 *h.)*

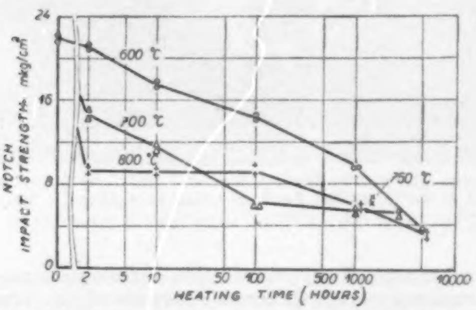


3 *Tensile strength and yield strength curves during the hardening of* AKRN *steel*

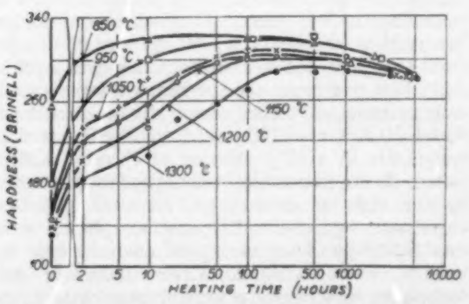
In the second stage of heat treatment, *i.e.* during precipitation heat treatment, considerable changes in hardness occur in relation to the temperature and time of heat treatment. In fig. 1 are shown the hardening curves for a melt containing 2·15% Ti. In consequence of the fairly high Ti content in this melt, hardening occurs over a wide range of temperatures. The hardening curves follow normal laws, for the rate of hardening decreases as a function of temperature, and the hardness maximum is displaced towards shorter periods as a function of temperature. The hardness maximum for this melt reaches a value of 320 H_B. In fig. 2 are shown the hardness values attained after a constant time of 10 h. and at different temperatures for three melts with different Ti contents. From this figure may be seen the relationship between the hardening sequence and the Ti content of the alloy; whereas for the melt containing 2·15% Ti at 800°C., a hardness very close to the maximum is reached, and at 850°C. signs of hardening are still apparent, the steel containing 1·05% Ti already remains completely unhardened at these temperatures. It is evident that with the decreasing Ti content the maximum attainable hardness is displaced towards lower temperatures. From this the conclusion follows that the hardening temperature should be adapted to the Ti content of the alloy.



4 *Elongation and reduction in area curves during the hardening of* AKRN *steel*



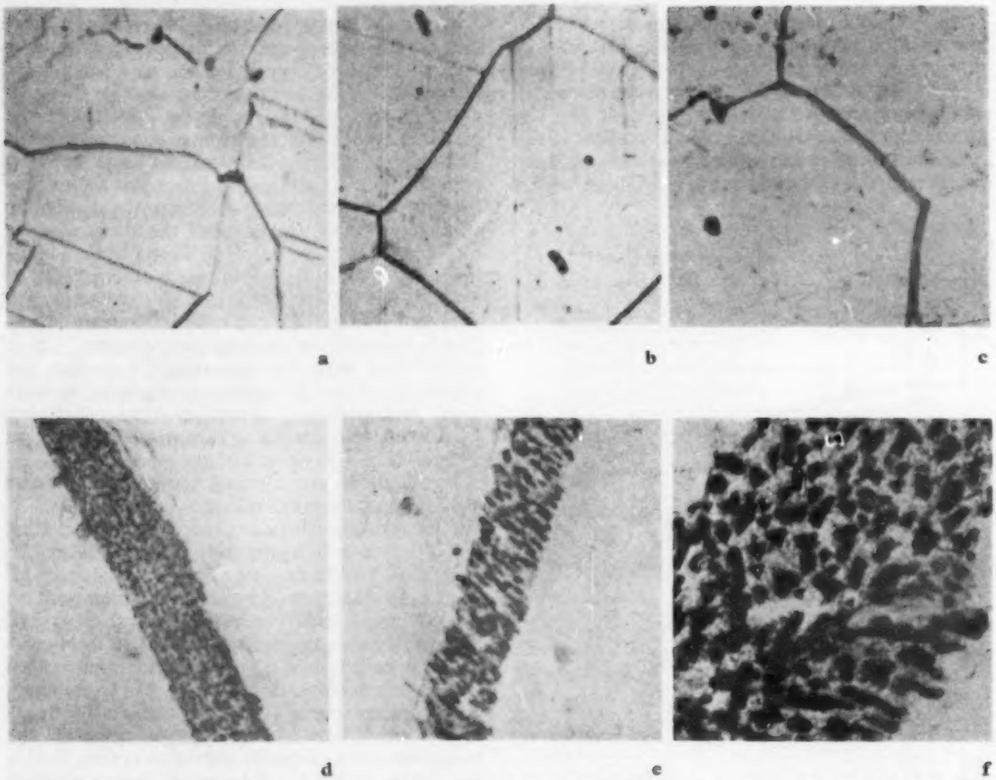
5 *Notch impact strength curves during the hardening of* AKRN *steel*



6 *Hardness curves for the hardening of* AKRN *steel at a temperature of 700°C. after solution heat treatment at temperatures of 850–1,300°C.*

Fig. 3 shows the ultimate tensile strength and yield strength curves for the melt containing 2·15% Ti. It is evident that during precipitation hardening the ultimate tensile strength increases from the original 62 to 108 kg./mm.² and the yield strength from 25 to more than 79 kg./mm.² By their position the maxima of these curves roughly correspond to the maximum on the hardness curves. The elongation and reduction in area fall in accordance with the increasing tensile strength (fig. 4). Overageing, which commences at 700°C. after about 100 h., is here characterized by the start of a fall in the elongation and a rise in the reduction in area. Events are similar at 800°C. The notch impact strength (fig. 5) decreases at all temperatures right from the start of precipitation heat treatment, and its initial drop is all the greater the higher is the temperature of this heat treatment. This decrease in notch impact strength is evidence of local reactions at the grain boundaries, since it is known that even a small quantity of a phase which is formed by local precipitation at the grain boundaries brings about considerable embrittlement.

The melt mentioned with the higher Ti content revealed an especially marked drop in notch impact strength, the values of which fell from the original 22 kg.m./cm.² after 5,000 h. heat treatment to about 4 kg.m./cm.² at all precipitation treatment temperatures. For melts with a lower Ti content the ultimate tensile strength and yield point attain lower maximum values after hardening, just as applied to the hardness. On the contrary, the elongation, reduction in area and notch impact strength have higher values. The melt with a low content of 1·04% Ti, for instance, has a maximum tensile strength of 69 kg./mm.² and a yield strength of 40 kg./mm.² with elongation 32% and reduction in area 54%, while the notch impact strength is



7 *Local precipitation of the carbide* Cr_7C_3 *at the grain boundaries after solution heat treatment at 1,150°C. for 2 h. with quenching in water;* **a, d:** *600°C., 2 h.;* **b, e:** *600°C., 1,000 h.;* **c, f:** *650°C., 1,000 h.;* **a, b, c:** *optical micrographs* × 500; **d, e, f:** *extraction replicas* × 10,000

11 kg.m./cm.^{2} For the melt containing 1·6% Ti the tensile strength after precipitation hardening is 92 kg./mm.^{2}, the yield strength 54 kg./mm.^{2}, the elongation 26%, the reduction in area 45% and the notch impact strength 9·5 kg.m./cm.^{2}

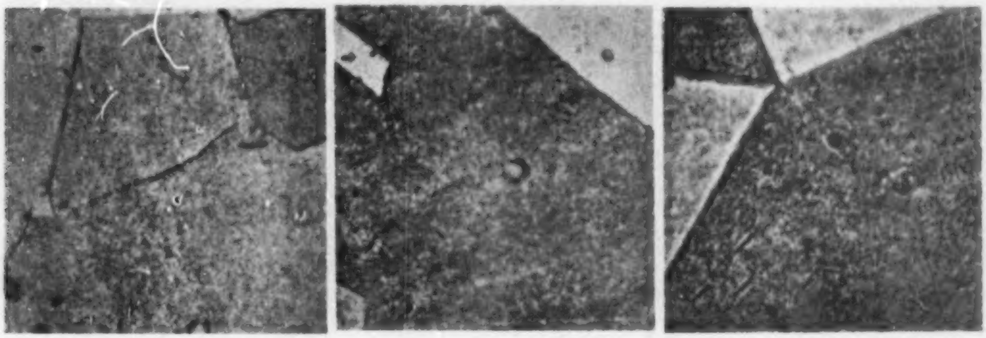
The results presented were obtained from specimens which were subjected to solution heat treatment at 1,150°C. for 2 h. with quenching in water before precipitation hardening. Similar batches of specimens showed, however, that the overall character of the curves obtained is unchanged if the solution heat treatment is carried out at any desired temperature between 1,050 and 1,200°C. (fig. 6).

Changes in structure during precipitation hardening

In order to investigate the changes taking place in the structure of AKRN steel during precipitation hardening, we employed optical and electron microscopy, X-ray and electron-microscope structural analyses and differential thermal analysis. The specimens for this part of the work were subjected to solution heat treatment at temperatures of 1,050, 1,150, 1,200 and 1,300°C. for a period of 2 h. and were quenched in water. They were precipitation hardened at 600, 650, 700, 750, 800 and 850°C. for from 1 to 1,000, and in some instances up to 5,000 h. After heat treatment, the specimens were mechanically polished and etched chemically in a mixture of 92 ml. HCl, 3 ml. HNO_3 and 5 ml. H_2SO_4, or electrolytically in a 10% solution of chromic acid in water. As the preparation technique for observation in the electron microscope we used collodion extraction replicas.[17] Individual phases were established by X-ray structural analysis with the use of mono-chromatic CrKα radiation, and namely on isolates which were obtained by electrolytic isolation in an electrolyte based on citric acid.[18] When it was a matter of analysis of extremely

a b c

8 *Precipitation of the fibrous titanium carbide (solution heat treatment at 1,150°C. for 2 h. with quenching in water):* **a, b:** *700°C., 200 h.;* **c:** *700°C., 1,000 h.; extraction replicas:* **a, b:** × 10,000; **c:** × 20,000



a b c

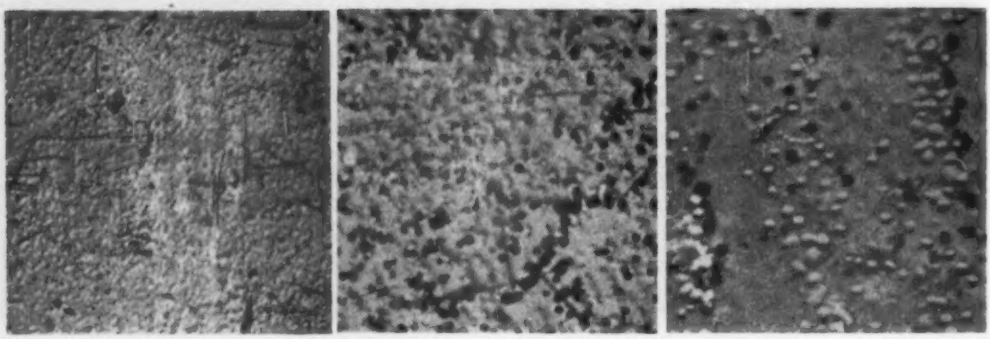
9 *Precipitation of the γ' phase:* **a:** *750°C., 200 h.;* **b:** *750°C., 1,000 h.;* **c:** *800°C., 5,000 h.; optical micrographs* × 500

fine phases visible on the electron micrographs, for their analysis we used electron diffraction on particles captured on the extraction replica, or a large number of these replicas were used for the preparation of a composite specimen for X-ray structural analysis. In certain instances, for comparison, analysis was carried out by electron diffraction of the fine particles directly on the surface of the metallographic specimen.

By these methods it was possible to detect in the structure several processes which we shall describe below.

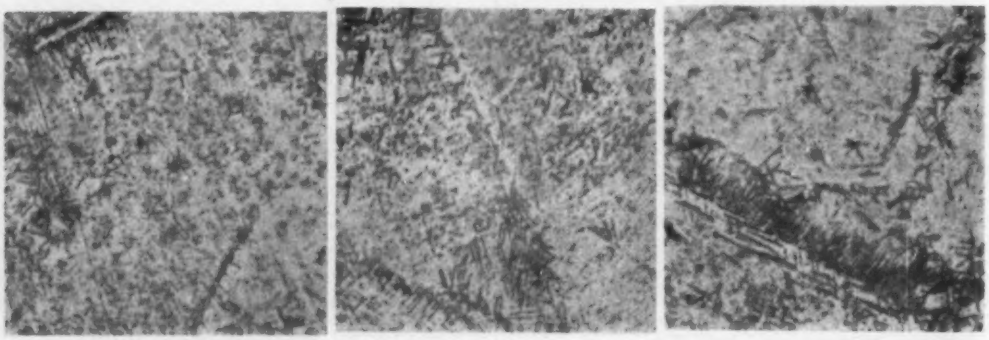
(*a*) *Precipitation of chromium carbide at the grain boundaries.* In figs. 7*a–f* it is shown that in the matrix of the specimens subjected to precipitation heat treatment at a temperature of 600–650°C. no visible precipitates are manifested. But even from the optical microstructures it is apparent that at the grain boundaries processes are occurring which are revealed by the formation of strips along the grain boundaries which have different etchability. The electron micrographs show that it is a matter of local precipitation of fine particles. These particles gradually increase in size with the increasing temperature and time of heat treatment and acquire characteristic shapes which are typical of chromium carbide. By structural analysis of these particles isolated from the matrix it was established that it was a matter of the carbide Cr_7C_3.

(*b*) *Precipitation of the fibrous titanium carbide.* At heating temperatures of 650–700°C., in addition to the particles at the grain boundaries, in the matrix there is the start of precipitation of fibrous precipitates, which at the outset are at the limit of the power of resolution of the electron microscope employed and later grow into visible needles (figs. 8*a–c*). In this later stage they are already accompanied also by a globular precipitate. By repeated electron diffraction on the extraction replicas and on the compounded specimens, it was



d e f

9 *Precipitation of the γ' phase:* **d:** 750°*C.*, 200 *h.*; **e:** 750°*C.*, 1,000 *h.*; **f:** 800°*C.*, 5,000 *h.*; *extraction replicas* × 10,000



a b c

10 *Precipitation of the η phase (solution heat treatment* 1,300°*C. for* 2 *h. with quenching in water):* **a:** 800°*C.*, 500 *h.*; **b:** 800°*C.*, 1,000 *h.*; **c:** 800°*C.*, 2,000 *h.* *Optical micrographs* × 500

correspondingly shown that the diffraction lines obtained can be attributed to the carbide TiC, or the carbonitride Ti(C,N), and the γ' phase. The relevant analysis is given in table 1. The precipitation of the carbide phase is not homogeneous within the whole of the matrix, since it starts preferentially in the vicinity of the grains of primary TiC carbides, and at other individual points, which were probably enriched in titanium during the solution heat treatment.

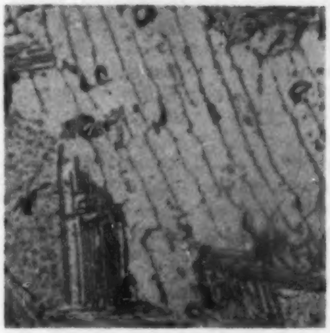
This result is in agreement with the morphological and structural findings of earlier authors who discovered a similar phase in austenitic, stainless steels stabilized with titanium or niobium.[19–21]

(*c*) *Precipitation of intermetallic compounds.* In the structure of the specimens heated at 700 and 750°C. a globular phase is manifested, the quantity and size of which increases with temperature and time (figs. 9*a–f*). By X-ray structural analysis of the isolates and extraction replicas it was consistently

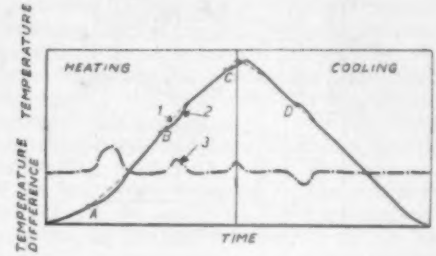
TABLE 1 *Structural analysis of particles on the extraction replica of the specimen in fig.* 8*c*

Interplanar distances, d Å		
Measured	Tabulated for	
	γ'-phase	TiC
2·52	2·54	
2·46		2·47
2·14		2·15
2·07	2·07	
1·81	1·80	
1·60	1·60	
1·52		1·52
1·46	1·46	
1·30		1·30
1·27	1·265	
1·24		1·24

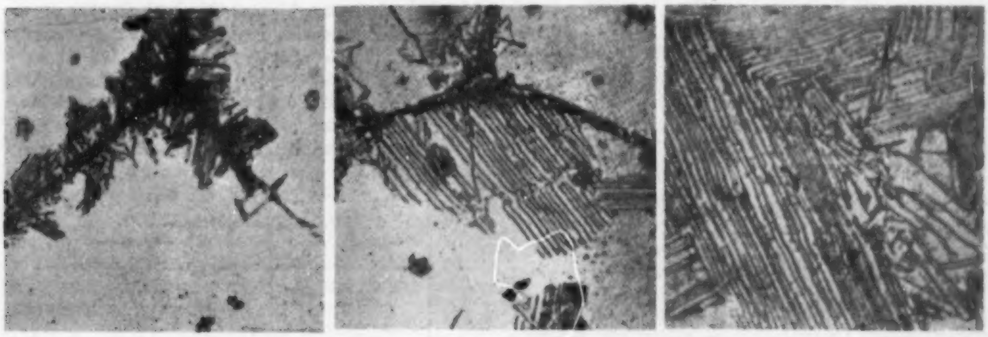
established that this is the γ' phase of composition $Ni_3(Al,Ti)$. The electron micrographs show that in a considerable proportion of the particles of the



11 *Recrystallization: Solution heat treatment: 1,050°C., 2 h., quenching in water; precipitation heat treatment: 800°C., 1,000 h. Optical micrograph* × 500



13 *Diagram of the sequence of thermal analysis*



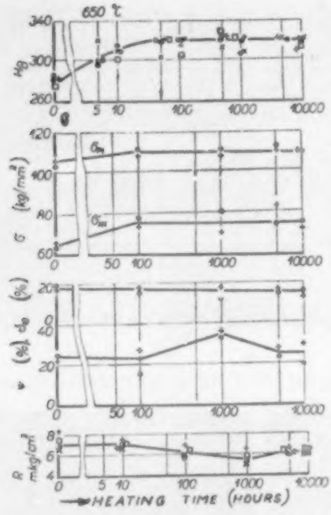
a b c

12 *Progress of recrystallization: Solution heat treatment: 1,300°C., 2 h., water; precipitation heat treatment:* **a:** *750°C., 5 h.;* **b:** *850°C., 100 h.;* **c:** *850°C., 1,000 h. Optical micrographs* × 500

γ' phase ordering along definite crystallographic planes is apparent. On the basis of this fact we judge it to be possible that the γ' phase is precipitated at this time stage preferentially in definite crystallographic planes in the shape of coherent platelets, which at a later stage after loss of coherence will break down into individual globules. The orientation of these globular particles then preserves the form of the original platelets, as is especially clearly shown by fig. 9*c*, on which the direction of the platelets is oriented at right angles to the plane of polish. We consider that this phenomenon is dependent on the kinetics of the precipitation of the γ' and η phases, as Soviet authors have mentioned.[22] We shall concern ourselves with the details of this phenomenon in a separate paper.

After long-term heating at higher temperatures above 750°C. in the advanced stage of overageing in the structure further phases are manifested in the form of lamellae distributed in certain crystallographic planes of the matrix (figs. 10*a–c*). This is the hexagonal η-phase having the composition Ni_3Ti.

(*d*) '*Recrystallization.*' During a prolonged period of ageing at temperatures over 800°C. a remarkable phenomenon occurs. This is characterized by the formation of a mixture of two phases, which is reminiscent of lamellar pearlite (fig. 11). By X-ray structural analysis it was established that it is a matter of an equilibrium γ and η phase. According to Geisler[23] it is possible to consider this discontinuous (microscopically inhomogeneous) precipitation as recrystallization, which leads to a reduction in the free energy of the stressed matrix. During normal precipitation, likewise, the free energy of the system as a whole, of course, decreases, but the free energy of the matrix remains high as a consequence of the plastic deformations brought about by the growth of the precipitate. As soon as the



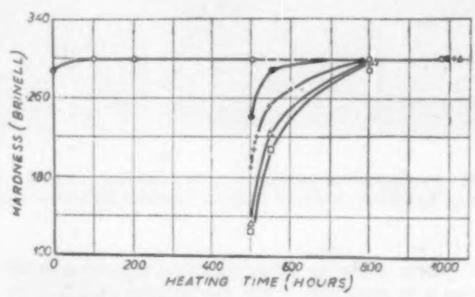
14 *Curves of the properties of* AKRN *steel during long-term heating at a temperature of 650°C.*

energy of the matrix attains a value sufficient for the nucleation of new grains which are devoid of plastic deformation the recrystallization process starts to take place. This event starts at the grain boundaries where the energy of the matrix is highest as a result of the primary local precipitation. The growth of the new grains of the lamellar mixture leads to a reduction in the free energy of the matrix and of the whole system, regardless of the fact that at the front of the growing grain dissolution of the precipitate deposited earlier must occur. If the specimen is at the relevant temperature for the required time, recrystallization extends to the whole volume of the matrix (figs. 12*a–c*).

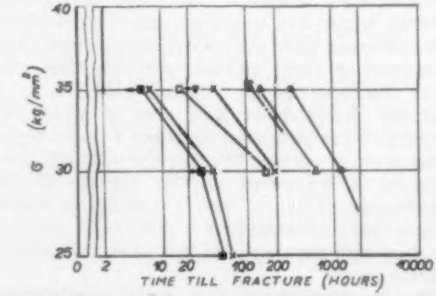
(*e*) *K structure.* Apart from the events which lead to visible changes in the structure, at low precipitation-treatment temperatures an arrangement of the atoms in the lattice takes place, linked with the formation of a so-called K structure. Detection of this phenomenon can be carried out by thermal analysis, which shows very sensitively all changes in the structure during heating and cooling. In our work, for differential thermal analysis, we used the method and equipment which have already been described earlier.[34]

The sequence of the changes was investigated on specimens hardened at 700°C. for 10 h. and at 750°C. for 25 h. after solution heat treatment at 1,050 and 1,150°C. for 2 h. with quenching in water. Heating for the thermal analysis was carried out at a rate of 10°C./min. and cooling at 6·7°C./min. Heating proceeded up to a temperature of 1,300°C., and after subsequent cooling additional cycles of thermal analysis were also carried out on the same specimens.

A plot of the recorded temperatures is shown in fig. 13. Curve 1 represents the plot of the temperature of the calibration standard, curve 2 the plot of the temperature of the specimen, and curve 3 gives the difference between the temperature of the standard and that of the specimen. During analysis of the curves obtained, it is, of course, necessary to bear in mind that during precipitation processes a very important part is played by time, so that the results of differential thermal analysis cannot present us with accurate information concerning the real temperatures of the start of the individual structural changes since these results are influenced by the heating and cooling rates. It is shown that the first endothermic change starts at 180–230°C. (point A), and this both for the specimens which were hardened (1st cycle of thermal analysis) and also for the specimens which were freely cooled (2nd cycle of thermal analysis). This change is characterized by the large quantity of heat absorbed, and its end was found with fairly considerable fluctuation around a temperature of 450°C.



15 *Hardness curves of* AKRN *steel after overheating at temperatures of* 800 *to* 950°C.



16 *Relationship between the time until fracture of* AKRN *steel and the conditions of overheating and renewed precipitation hardening without solution heat treatment*

In this connection, it is necessary to point out that in Nimonic alloys a pre-precipitation stage was found, which was linked with the formation of Guinier-Preston zones, just in the same way as in light alloys.[25] Livshits *et al.*[26] showed that in these alloys after gradual cooling there occurs a transformation of the disordered solid solution existing at high temperatures into an ordered solid solution, named by Thomas the K structure. During heating, on the contrary, transition of the ordered structure into a disordered one takes place, which is linked with heat absorption. But if a specimen is rapidly cooled from the solution heat-treatment temperature, the disordered solid solution remains preserved, and during heating formation of a K structure (liberation of heat) takes place first of all, and then during subsequent heating the transformation to a disordered solid solution (heat absorption) once again occurs. This explanation can also be accepted for clarification of the process which takes place in the steel under study during heating at low temperatures.

This was also confirmed by a further experiment, during which differential thermal analysis was used on specimens which were quenched in water after two hours' solution heat treatment at a temperature of 1,050 or 1,150°C. On the relevant curves an exothermic reaction was first of all revealed, characteristic of the formation of a K structure, after which there immediately followed an endothermic reaction, thus manifesting the formation of a disordered solution.

Relationship between structural changes and mechanical properties

From a comparison of the changes in the mechanical properties during hardening with the changes observed in the structure certain relationships follow. In the first period of precipitation hardening, when the hardness, tensile strength and yield point increase, we do not observe any changes in the matrix even with the use of an electron microscope with a power of resolution of about 100 Å. It may be suggested that the hardening is a result of the effect of coherent formations of hardening phases, which by the methods used are visible until after the loss of coherence with the basic solid solution. We observe formation of the γ' phase on the replicas roughly at the point where the hardness, tensile strength and yield point curves pass through maxima. Formations of the η-phase are manifested in the structure long after the hardness maximum is reached; for instance, after solution heat treatment at 1,150°C. for 2 h. and quenching in water and precipitation hardening at 800°C. this phase is manifested even after 5,000 h.

On this basis we judge that the main hardening factor should be considered to be the precipitation phase γ'-$Ni_3(Al,Ti)$. Even though from the experiments carried out it is impossible to provide convincing proof of how far the mechanical properties of the steel are influenced by this phase in the period which precedes its resolution by the electron microscope, it may be assumed that its action is not marked in this instance since, even in this period, a substantial reduction in the resistance of the steel to deformation takes place. The favourable action of the increasing Ti content on the precipitation hardening of the steel would them imply above all an effect on the lattice parameter of the phase $Ni_3(Al,Ti)$ in the direction of a greater difference between its parameters and that of the basic solid solution and thereby, too, an increase in the hardening action of the coherent γ' phase. Likewise precipitation of the η phase, being a product of recrystallization, is not manifested in the course of the hardening curves.

Concerning the fibrous precipitate Ti(C,N), it is difficult to establish to what extent it participates in the hardening of the matrix since it is precipitated simultaneously with the γ' phase. But it may be assumed that it does not play a special part owing to the fact that it is not precipitated homogeneously, but predominantly at points enriched in titanium.

The continuous fall in the notch impact strength from the very start of the precipitation heat treatment must be attributed to local precipitation of carbide phases at the grain boundaries.

Changes in mechanical properties during long-term heating at operational temperature

During precipitation hardening in accordance with the production technique, the cycle is kept as short as possible on economic grounds, and maximum hardness values are not reached. Precipitation hardening also proceeds further in service, and sooner or later, according to the temperature level, a maximum is reached which is dependent on the titanium content. Jointly with the hardness, there is an increase in the tensile strength and yield strength. At an operational temperature of 650°C. the steel under test reaches maximum hardness, tensile strength and yield strength after a period of 10,000 h., and neither the elongation nor the reduction in area are changed (fig. 14). The notch impact strength shows a moderate drop, which is a sign of the structural changes taking place at these operational temperatures, *i.e.* mainly continuing precipitation at the grain boundaries. In the instance mentioned, the drop in notch impact strength is small, since as a result of the high Ti content (2·15%) the initial values are already low. In the melt containing 1·05% Ti this drop is more marked, for after 5,000 h. at a temperature of 650°C. the notch impact strength drops from 11 to 8 kg.m./cm.² It may be judged that the steel investi-

gated is sufficiently stable during long-term service at a temperature of 650°C. and still has satisfactory notch impact strength values. At higher temperatures overageing takes place, which is linked with a reduction in the resistance to deformation. At 700°C. there is a drop in hardness even after 100 h.

Changes in properties after overheating

If, in service, a short-term increase in the temperature takes place (during sudden overheating or on technological grounds, such as brazing or welding) a reduction in hardness and an increase in notch impact strength of the tested steel ensue, as evidence that partial dissolution of the hardening phase took place. During subsequent heating at the operational temperature precipitation hardening again takes place, the rate of which is dependent on the excess heating temperature and the temperature of the new precipitation hardening.

In order to establish the time required for complete precipitation hardening of overheated specimens we proceeded as follows. A specimen originally hardened (1,150°C./2 h./water—780°C., 10 h. —750°C., 25 h.) was held for the first 500 h. at a temperature of 650°C. Within this period additional precipitation hardening took place, and then the hardness remained constant. After 500 h. the specimens were artificially overheated at 850°C. for 5 h. and cooled in air, so that a substantial reduction in hardness and increase in notch impact strength resulted. During subsequent heating at the operational temperature of 650°C. the specimens regained the original hardness after 300 h. If after this time further overheating follows, the specimen can again be hardened at 650°C. to the original hardness. Exactly the same thing happened during a further, third cycle.

This original hardness can be attained after 300 h. at a temperature of 650°C. even on specimens overheated for 5 h. at a temperature of 950°C. (fig. 15).

By structural analysis it was found that during overheating of hardened (or re-aged) specimens dissolution of small particles of the γ' phase takes place with simultaneous growth of the large, more stable particles. During subsequent treatment at the low (operational) temperature, hardening occurs as a result of precipitation of the γ' phase, which is partly made possible by the preceding dissolution of the γ' particles, but primarily implies further impoverishment of the solid solution in hardening components. Evidence of this was especially provided by the results of tests whereby it was established whether on overheated material the original creep tensile strength is again reached simultaneously with the original hardness. For this purpose, a number of tests were carried out at 650°C. and a stress of 35·30 and 25 kg./mm.^{2} This was done both on specimens hardened at 750°C. for 10 h. and overheated for 5 h. at temperatures of 850, 900 and 950°C., on the one hand without further hardening, and on the other hand hardened at a temperature of 650°C. for a period of 300 h., *i.e.* to the original hardness.

The results of the experiments are shown in fig. 16. It is shown that specimens overheated and again hardened collectively reveal shorter periods until fracture than specimens not overheated, in evidence of the fact that by overheating and renewed hardening without solution heat treatment the basic solid solution is impoverished in the hardening component, the precipitation of which in the normal instance favourably influences the creep resistance at high temperatures.

The continuous drop in the notch impact strength from the very start of precipitation heat treatment must be attributed to local precipitation of carbide phases at the grain boundaries.

Conclusions

Structural and mechanical tests on hardenable, creep-resistant 35Ni-15Cr-3W-Ti,Al show that the hardening of this steel is linked primarily with the precipitation of the γ' phase of composition $Ni_3(Ti,Al)$. Another intermetallic phase of composition Ni_3Ti is manifested in the structure at an advanced stage of overageing, and its formation is not manifested on the curves expressing resistance to deformation. In the time period of precipitation, in the structure particles are manifested of the fibrous carbide or carbo-nitride Ti(C,N), which is precipitated primarily in areas enriched in titanium. At the grain boundaries local precipitation of the chromium carbide, Cr_7C_3, takes place.

By tests on overheated specimens it was once again confirmed that the high hardness of hardenable alloys still does not ensure their good creep resistance. Overheated specimens, which have again been hardened without further solution heat treatment, will indeed reach the same hardness as specimens which have not been overheated, but have a low creep tensile strength, since by this process the basic solid solution is impoverished in the hardening component.

The good structural stability of this steel at temperatures around 650°C. for which it is primarily intended, permits its long-term use for the purposes of stationary heat engines.

Acknowledgments

It is our pleasant task to thank P. Schier from the Metallurgical Institute of the Czechoslovak Academy of Sciences for making possible work on the electron microscope and J. Ševčík from this institute for help during this work.

References will be given next month

Russian forging journal

Abstracts from the Russian forging journal—'Kuznechno - Shtampovochnoe Proizvodstvo,' March, 1961, **3.** *This is the third year of this journal devoted specifically to forging. Short abstracts of the more important articles are given in* METAL TREATMENT *each month.*

Effect of the deformation conditions and of the subsequent heat treatment on the properties of EI 437*B alloy.* M. YA. DZUGUTOV and B. F. VAKHTANOV. Pp. 3–7.
The highest and most stable properties of this heat-resistant steel are obtained after heating at 1,080°C. for 8 h. and at 750°C. for 16 h. Best mechanical properties are obtained after deformation between 1,000 and 1,100°C. and the greatest creep resistance after deformation at 1,160°C.

Calculation of the temperature of the flash during hot stamping on presses. E. A. SONKIN. Pp. 8–9.
Analytical calculation of this temperature makes it possible to carry out even more confidently the development of technological processes of hot stamping on presses and to design the dies. Calculations of the temperature of the flash also make it possible to throw light again on problems of the effect of its thickness on the choice of the power of presses and on the working life of the dies.

The use of a new steel for blanking dies to increase their life. V. I. ZALESSKII, F. P. MIKHALENKO and V. V. GUBAREV. Pp. 9–16.
Tests are described of the working lives of dies made of various steels.

Some problems of vibration working of metal with restriction of speed. M. YA. KARNOV. Pp. 16–18.
A comparison is made of the results of forging on vibration and conventional presses, and of their different effects on structure and mechanical properties.

Increasing the accuracy of the determination of the force required for the compression of forgings. E. N. MOSHNIN and N. M. ZOLOTUKHIN. Pp. 18–19.
With reference to the authors' earlier article (*Kuznechno-Shtampovochnoe Proizvodstvo*, 1960, **2**, No. 6), corrections are now suggested to increase the accuracy of determination of the mean temperature of a forging of over 15 tons, where the cooling conditions are no longer regular, and the temperature of the centre of the forging remains relatively constant. Correction factor values are given for the equations derived in the earlier article. Care should also be taken in the measurement of surface temperatures with optical pyrometers, since the presence of even a slight layer of scale at the point of measurement may lead to an error of up to 150°C. in the temperature value recorded.

Forging and stamping cast iron. S. I. KLYUCHNIKOV. Pp. 19–22.
Experiments described show that white cast iron can be forged to manufacture tools and other components in the temperature range 1,050–850°C. In view of the limited capacity to compress and draw the material in any one pass, the use of repeated heating periods is essential. It is expedient to use special forging cast irons for the manufacture of components of complicated outline, when accurate dimensions can be obtained and the annealing cycle after machining can be shortened. It is a feasible possibility to produce pinions, discs and other hot stampings, especially from S.G. high-strength irons. The method of liquid stamping may also be employed, and this produces an even greater improvement in mechanical properties than ordinary forging and stamping. (Liquid stamping, a process developed in the USSR, consists of four stages: gravity pouring of the metal into an open metal mould; pressing of the liquid metal during which shaping and solidification take place; holding of the mechanically worked metal in the die, initially used as the mould, when the ram is gripped by the metal due to shrinkage of the latter; and withdrawal of the finished components from the die or mould, when it is stripped from the ram. Abstractor.)

Determination of the dynamic loads on the camshaft mechanisms of vibratory-type cold upsetting machines. YU. A. MIROPOL'SKII. Pp. 23–26.
In a similar manner to the author's earlier article (*Kuznechno-Shtampovochnoe Proizvodstvo*, 1960, No. 8) calculation methods are derived from cold upsetting machines, whose pusher arms carry out a vibratory movement.

Multi-position forging presses. L. A. PRISHCHEPIONOK. Pp. 27–30.
Twenty 400-ton presses built by the Barnaul'sk Works are described, with dimensions and operating characteristics.

Determination of the loads on the components of the

continued on page **444**

LETTERS *to the Editor*

Note on scaling of steel

SIR: In the article, 'Observations concerning the structure of scale, and the mechanism of formation during the heating of steel,' by Sten Modin and Erik Tholander (METAL TREATMENT, July, 1961, from the original Swedish, *Jernkontorets Annaler*, November, 1960), the authors state on p. 262 that: 'No gas analyses were made in relation to the furnace atmosphere.' This was unfortunate for it would seem that an opportunity was missed to relate the different scaling characteristics in an oil-fired furnace and in the atmospheric air, electric furnace to the respective differences in oxidizing potential of the two atmospheres.

It is likely that the products of combustion from the oil firing would contain certain sulphur compounds (SO_2 or H_2S, depending on the oxidizing potential) which would undoubtedly influence the nature of scaling. It is well known that the presence of complex, low melting-point, oxide/sulphide eutectics at the metal/scale interface can cause local fusion within the scale leading to the formation of bright, shiny areas as indicated in figs. 3*a* and 3*b* on p. 264.

The varied appearances of the scale produced by heating in an oil-fired furnace would seem to be related to the oxygen/sulphur concentration at the steel surfaces, such that a low oxygen potential would give a sparkling, crystalline surface (*cf.* fig. 3*b*, p. 264, representing a test-piece placed on the side of the furnace opposite the burner) and a higher oxygen potential would give a shiny, almost smooth surface (*cf.* fig. 3*a*, p. 264, representing a test-piece placed on the burner side of the furnace).

Recent unpublished work by the writer, on the relationship between furnace atmosphere and sulphur content of scales formed on steel, tends to confirm the above opinion.

R. ROLLS

The Manchester College of Science Technology,
Department of Chemical Engineering,
Fuel Technology and Metallurgy
September, 1961

Author's reply

I have read Mr. Rolls' remarks with great interest and I can answer that experiments with varied furnace atmospheres are planned by us but it has not yet been possible to get them started. Mr. Modin and I have stated in our first report that further experiments were planned and will be reported on later as the results are available.

It will be of great interest to know the effect of the sulphur content in the furnace atmosphere on the scale appearance, which Mr. Rolls writes about.

ERIK THOLANDER

Sveriges Mekanförbund,
Eskilstuna, Sweden
September, 1961

Russian forging journal

concluded from page 443

clamping mechanism of a forging manipulator. V. G. MIRONOV. Pp. 30–32.
Research into these loads leads to recommendations for their reduction by improved design.

Choice of the source of energy for heating forgings in forge and stamp shops. (An assessment.) V. N. GLUSHKOV. Pp. 33–35.

A combined hydraulic press installation. P. D. MOKHOV, I. F. KAPLENKOV and S. F. MARGOLIN. Pp. 35–36.
The 600-ton hydraulic press installation for producing automobile frame members at the Minsk automobile works is described.

New designs of blanking dies for forgings produced on crank-drive hot forging presses. D. E. SHAPOSHNIKOV. Pp. 36–38.

Stamped separators of roller guides. E. M. GIL'DINSON and A. M. GRUNTOV. Pp. 38–39.

A universal holder for rapidly changeable dies. V. M. KOLOSOV. Pp. 39–41.

Piercing of cylindrical apertures in forgings with bosses/hubs. S. V. LYASHENKO. Pp. 41–42.

Modernization of a forging manipulator. YU. A. ZHURAVLEV. Pp. 42–43.

A new design of packing for pistons in working cylinders of air-steam hammers. P. D. NEPECHII and S. A. VOL'SKII. P. 44.

Study of the die-forging operation in presses

A LARGE FACTORY, specializing in the production of spoons and forks, has carried out numerous tests on the dynamics of the pressing operation.[1, 2]

Measurement of pressures

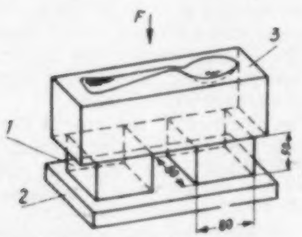
The method of measuring used employs resistance gauges. Fig. 1 is a diagram showing how the measuring device is mounted on the press. The bottom die rests on two steel blocks placed on the bedplate of the press. All the surfaces in contact have been carefully ground to ensure uniform distribution of forces. On the vertical sides, standard resistance gauges have been attached with bridge connection.

In another variant the standard gauges are replaced by a continuous *ad hoc* coil 80 × 80 mm. ($3\frac{1}{8}$ × $3\frac{1}{8}$ in.) of Constantin wire 0·05 mm. (0·002 in.) dia. The coil is insulated from the block by means of ceramic beads placed at the vertical corners. The advantage of this arrangement is that it allows the effective contraction of the block to be measured even in eccentric positions of the pressure centre.

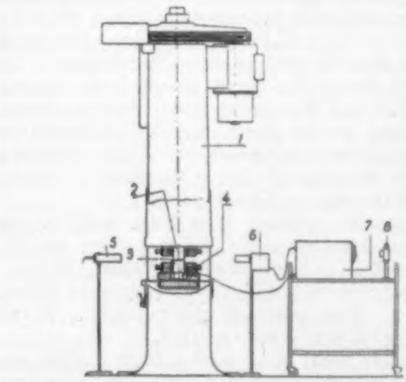
In fig. 2 is shown the complete measuring set-up for a Weingarten screw-press (stroke 245 mm.). The out-of-balance current in the measuring bridge is amplified and fed to a cathode-ray oscilloscope. The switching on of the current is controlled by a photoelectric cell at the moment of contact of the upper tool with the work-piece. The oscilloscope reading is recorded on film at mains frequency (50 c./s.). Each of the two blows required to produce a proper shape of the work-piece, *i.e.* roughing and finishing, are recorded. On the press mentioned above, with a stroke of 245 mm., these were: roughing stroke, 571 tons; finishing stroke, 587 tons.

This is a rather surprising result, since the noise of the finishing stroke being definitely louder, it was expected that there would be a markedly large difference in the pressure.

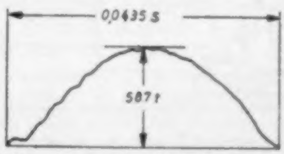
As regards the distribution of the forces between the two blocks, the recordings showed 40 tons



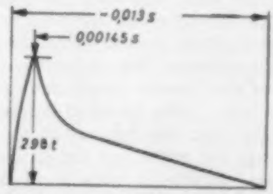
1 *Arrangement of measuring blocks*
1, Measuring block. 2, Base plate. 3, Bottom die



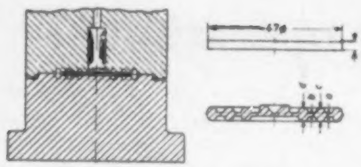
2 *General arrangement of measuring apparatus*
1, Press. 2, Top die. 3, Bottom die. 4, Measuring block 5, Light source. 6, Photo-electric cell. 7, Cathode-ray oscilloscope. 8, Camera



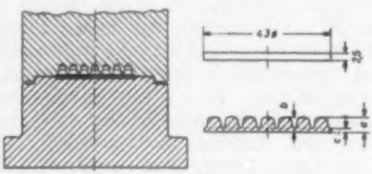
3 *First striking of spoon in nickel silver on press. Stress-time*



4 *Drop forging of spoon in chromium nickel steel under hammer (handle end). Stress-time*



5 *Tools for tests on relief No. 1.*
(*Letters indicate the thicknesses measured*)



6 *Tools for tests on relief No. 2.*
(*Letters show thicknesses measured*)

more on the handle-end block than on the block at the other end. The profile of the relief has an influence on the value of the maximum pressure, which can, in extreme cases, reach a value of 750 tons.

In later investigations, only one measuring block has been used, with dimensions of 280 × 80 × 80 mm. (11 × 3⅛ × 3⅛ in.), and the second channel has been used for reading mains fluctuations.

Fig. 3 shows the curve recorded for forging a table fork on a Weingarten press, 245 mm. stroke. The ripple on the curve must be attributed to vibrations of the press framework. These vibrations are more pronounced during the finishing stroke in which the effect of damping is less.

Comparative readings have been made on an American compressed-air hammer (ram weight, 405 kg. (893 lb.); height of drop, 650 mm. (25⅝ in.). These gave the very different looking curve shown in fig. 4. The extremely short duration of the moment of impact is very marked.

When the metal being used is fairly ductile, the handle-end is decorated to a greater or less extent in relief and the stress on this end is greater. On the other hand, spoons and forks in chromium nickel stainless steel have little decoration and, in this case, the formation of the bowl or the prongs gives rise to a greater pressure than the handle end.

It can be assumed that the speed of the blow in the case of the hammer results in quicker destruction of the die. The required pressure depends on several factors: the delicacy of the embossing, the quality of the metal and the surface condition (roughness). Hydraulic presses of the order of 1,000 tons, which can be used for testing tools, are seen in works. The average stresses for spoons and forks in ductile metal are of the order of 600 tons, while for stainless steel a figure of 650 tons can be taken.

Influence of the relief

Tests made on this aspect have been carried out with a toggle press in the following way. Two disc dies were made (figs. 5 and 6) with which rings 2·5 mm. thick were struck. The materials used were: (*a*) brass, 63%; (*b*) nickel silver, 63–67% Cu, 11–13% Ni; (*c*) chromium nickel steel 18/8, 0·07% C.

To obviate the effect of variations in thickness, rings were selected to a tolerance of 0·01 mm.

The brass rings were subjected to ordinary annealing to reduce their Vickers hardness from 90–70 kg./mm. The nickel-silver rings were annealed in a neutral atmosphere; hardness before annealing, 100 kg./mm. The steel rings had a hardness of 147–160 kg./mm.

In the first case (fig. 5), the sunk surface was equivalent to 45% of the whole surface, whilst in the second case (fig. 6), it was 50%. It will be noticed that, in the second case, the top tool has 18% taper holes 5 mm. at base, of sufficient depth to leave the upper face of the deformed metal free.

Figs. 5 and 6 show the points at which measurements were taken by a comparator to an accuracy of about 1/100 mm. The force applied was determined by the method already described.

Pressures varied between the following limits:

	tons
Hydraulic press	150–500
Screw press	150–300
Toggle press	200–500
Hammer	300–600

The results of these tests have been plotted with stresses as abscissa and, as ordinates, the changes in dimension with reference to the nominal thickness of the original ring. The thickness at the edge of test-piece (fig. 6) as being dependent on the size of the flash was not taken.

From an analysis of the graphs obtained, the results show that with the same pressure maximum deformation takes place with the toggle press

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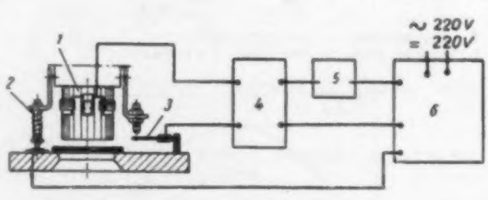


7 *Measuring of cutting out under press*
1, *Punch with tensionmeter gauge.* 2, *Contact of oscilloscope circuit.* 3, *Deflection beam with gauges for recording movement of punch.* 4, *Commutator.* 5, *Amplifier.* 6, *Coil oscillator*

Evaluating high-strength materials

This interesting discussion on the problems of deciding what factors count in choosing high-strength materials was given in ' Metal Progress,' September, 1961, by Mr. G. K. Manning, technical manager, Department of Metallurgy, Battelle Memorial Institute

IN RECENT YEARS, engineers have become increasingly aware that there is a weight penalty associated with every moving part. For example, it is well known that the weight penalty in missiles is enormous, and that associated with aircraft, very large. But there is also a weight penalty involved with more prosaic items such as automobiles, agricultural equipment, and even ships. Hence, the trend toward lighter and lighter weight, higher and higher service stresses and, perforce, materials of higher and higher strength.

The materials engineer may with some justification feel he is in a ' rat race. ' He no sooner meets one demand for strength with ' adequate ductility ' than he is confronted with a new demand for still greater strength—yet ' adequate ductility ' must remain unimpaired. Perhaps we should pause and take stock. The past may hold lessons that are applicable to the future.

In 1845 wrought iron was the strongest structural material being used for applications requiring reliability in the presence of tensile stress. It had a strength of about 45,000 lb./sq. in. By 1880, less-conservative engineers were building bigger and faster locomotives with medium-carbon steels which, when normalized, had strengths as great as 75,000 lb./sq. in. This was not an easy task. These ' strong ' steels were treacherous. They ' crystallized, ' particularly in cold weather. Since failures of piston rods, drive links and axles were numerous, many engineers concluded that these ' new-fangled ' steels were not practical. Yet, by the end of World War I, we were making and using some components in the Liberty engine that had a strength of 125,000 lb./sq. in., and these parts were subject to tensile stresses.

By World War II a number of aircraft frame and engine components were made from steels that required 190,000 lb./sq. in. tensile strengths. And today we are using steels that possess strengths as great as 275,000 lb./sq. in. Yet the engineer is still challenged to find stronger and stronger materials. How is he to do this?

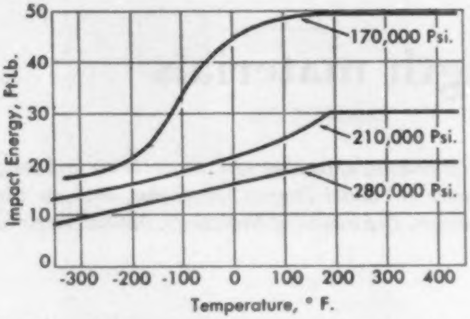
What is ' adequate ductility '?

Let us first discuss some factors that the engineer must consider. It is apparent that our concept of ' adequate ductility ' has been changing, particularly during the last 50 years. At the time of World War I, metallurgists knew how to heat-treat steels to a strength of 250,000 lb./sq. in.; these steels possessed almost as much toughness as present-day steels with the same strength.

Obviously, progress did not come about because the materials engineer was able to achieve ever-increasing strength while still maintaining a fixed degree of ductility. It is true that the metallurgist has been able to improve mechanical properties to some degree, but, equally important, the designer has found ways of avoiding stress concentration and the fabricator has found ways of detecting and eliminating flaws in the finished part. Through the years we have learned how to use increasingly brittle materials. Thus the phenomenal progress of past years has been a combination of (*a*) improved materials, (*b*) improved design and (*c*) greater perfection in the finished part. As a consequence of these advances, a given material no longer needs the ductility it needed in the past to be ' adequate ' for the particular application.

As another factor, the concept of transition temperature has become deeply embedded in our thinking within the last 15 or 20 years. A very helpful tool for evaluating ship plate, heavy forgings, armour plate and a variety of steels quenched and tempered to a strength of 200,000 lb./sq. in. or less, it will continue to prove useful for these materials for many years. However, as we move to still higher-strength levels, the transition-temperature concept places real limitations on prospects for progress.

Fig. 1 illustrates how impact strengths change as tensile strengths are increased. Not only does the transition temperature rise with increasing strength, but, more importantly, the energy value of the upper plateau decreases with increasing strength. Thus, at high strength levels, there is less difference



1 *Effect of different tensile strengths on impact strengths at various temperatures. As the tensile strength rises, the transition temperature rises and the energy value of the upper plateau drops. The steel is AISI* 4340, *oil-quenched and tempered.*
(*L. J. Klinger, W. J. Barnett, R. P. Frohmberg and A. R. Troiano*, Transactions A.S.M., 1954, **46**, *p.* 1557)

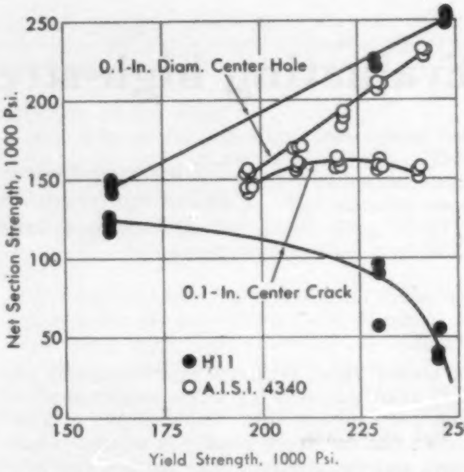
between ' brittle ' and ' ductile ' fracture in terms of energy absorbed and in terms of deformation prior to fracture. In other words, the transition temperature has less significance at a strength of 280,000 lb./sq. in than at a strength of 170,000 lb./sq. in.

Limitations of test methods

In addition, some recent experiments have shown that the load-carrying abilities of some notched sheet tensile specimens change in an anomalous manner with the notch severity. Fig. 2 shows data obtained on sheet tensile specimens containing centrally located defects of two different degrees of severity. Some of the specimens contained drilled holes 0·1 in. in diameter that produced nominal stress concentrations of about three. Other specimens contained centrally located fatigue cracks 0·1 in. long. In these specimens, the stress concentration prior to the onset of plastic deformation must have been several times greater.

It is not at all suprising, therefore, that the fatigue-cracked specimens failed at a lower load than did those containing a drilled hole. However, it is surprising that, for a given yield strength, the fatigue-cracked specimens of AISI 4340 were able to carry greater loads than similar specimens of H11, but that the H11 specimens with centrally located holes were stronger than AISI 4340 specimens with the same geometry. Apparently, the order of merit which one might assign to the steel depends to some extent on the degree of stress concentration.

Another way of varying stress concentration is to vary the length of the fatigue crack, the stress concentration prior to onset of plastic deformation being proportional to the square root of the crack length. Fig. 3 shows a group of H11 and AISI 4340



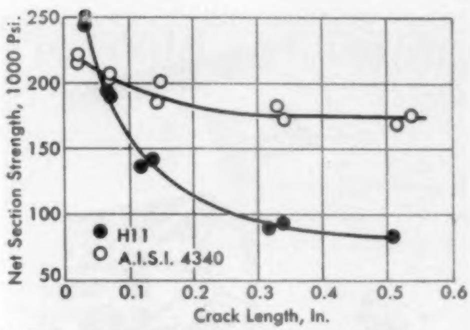
2 *Tensile data for sheet specimens containing either holes or fatigue cracks, centrally located. Note that the H*11 *specimens with holes appear stronger than the AISI* 4340 *specimens with holes, but that the reverse is true for H*11 *and AISI* 4340 *specimens which contain cracks. Stress concentration is therefore an important factor to consider in testing metals and alloys*

specimens quenched and tempered to a yield strength of 225,000 lb./sq. in., but containing fatigue cracks of various lengths. As is apparent, the curves for the two steels cross at a crack length of about 0·05 in. When the cracks are shorter than 0·05 in., H11 will sustain the greater load, but when they are longer than 0·05 in., AISI 4340 will sustain the greater load.

Figs. 2 and 3 are neither an endorsement nor an indictment of either AISI 4340 or H11. That is not the point. The point is that there are subtle differences in steels which the usual evaluation tests will not disclose. Thus, though AISI 4340 may seem to be superior to H11 on the basis of either V-notch Charpy tests or notched sheet tensile tests, figs. 2 and 3 suggest that such a generalization is not justified.

A suggested approach

For the engineer interested in building stronger and lighter structures there is, however, a rational procedure for evaluating materials—one that is consistent with both our present state of knowledge and with the teachings of history. It is necessary to realize that the load-carrying ability of a structure, at least a high-strength structure, is closely related to the stress concentrations which are present. There are two primary sources for these concentrations. One is a matter of how the structure is designed (changes in section size, and the like); the other, in advertent flaws (such as small cracks



3 *Effect of centrally located fatigue cracks of various lengths on net section strength of H*11 *and AISI* 4340. *Note that H*11 *appears stronger when cracks are shorter than* 0·05 *in., but that the reverse is true when the cracks are longer*

formed during fabrication, or non-metallic inclusions formed during casting).

Therefore, it seems reasonable to evaluate high-strength materials by measuring their load-carrying abilities in the presence of the maximum stress concentration likely to exist on the finished part. In some instances, for example, it may be appropriate to use a tensile test with a carefully filleted reduced diameter (when the maximum stress concentration is due to design). More frequently, however, some kind of sharp crack located perpendicular to the applied load will represent the most severe stress condition likely to be present in the finished part. For example, annular fatigue cracks located midway along the length of a standard tensile bar might be used to simulate internal flaws that could not be detected in massive structures. Thumbnail fatigue cracks in sheet tensile specimens might be used to simulate the most severe condition that exists in a solid-propellant motor case. Materials, then, could be rated on the basis of their load-carrying ability (net section strength) in the presence of the stress raiser that seems most appropriate.

Such a procedure would not always be simple and straightforward. To apply it, the engineer would need to determine the type of defect that might exist in the finished part and what the maximum size might be. Second, he would have to realize that the maximum size of defects probably would decrease with improvements in design, fabrication processes and non-destructive testing technique. Consequently, the kind of tensile test employed would have to be modified from time to time.

On the other hand, such a procedure would have two important advantages. First, if properly executed, it would produce data that could be directly applied to design. Second, it would focus attention on the very important effect that flaws have on the load-carrying ability of high-strength structures.

To sum up, past progress has been the result of simultaneously improving materials, design and fabrication techniques. It should be kept in mind that future progress probably will be the result of joint efforts by the materials engineer, the designer and the manufacturer.

Study of the die-forging operation in presses

concluded from page 446

whilst the minimum is obtained from using the hammer. The stress calculated for the effective relieved surface of specimen No. 1 (fig. 5), under a total pressure of 300 tons, is 427 kg./mm.^{2} Beyond 300 tons, the effect of deformation becomes insignificant.

In respect of the tests on specimen No. 2, the limit of usefulness of toggle and hydraulic presses is around 400 tons, and this corresponds to a stress of the order of 553 kg./mm.^{2}

Conclusions

These results give only some indication of the behaviour of the presses. They do not in any way solve the general question of die forging, in which other factors must be taken into consideration when the choice of manufacturing plant is being considered and a search is being made for tooling equipment capable of ensuring long life and a high return.

It is thought that this method and its further development might well be of service not only in studying the very delicate problem of forging with quick-acting presses but also in reducing the percentage of rejects in the trimming operation, where too rapid wear of the tooling complicates and hampers the final stroke. The method is also indicated in coining operations where formers are expensively produced from reduced reproductions of designs on blocks of very hard steel for striking medals and particularly coins.

Fig. 7 shows an apparatus for two-way measurement, one for recording pressures and the other for giving a curve of the displacement of the punch as a function of time.

The cleanness of the profile also depends on the play between the punch and the die as well as on the number of blows of the press and, consequently, on the speed of deformation of the metal.

References

(1) Ernst Fr. Scheren, *Zeitschrift für Metallkunde*, 1959, (2), 106.
(2) B. Puszet, *La Pratique des Industries Mecaniques*, 1960, (9), 251.

Solar furnace for high-temperature research

STANFORD RESEARCH INSTITUTE in California has recently constructed and put into operation a 'solar furnace.' This is the only one of its type operating in America, and will enable the research staff of the Institute to conduct certain experiments with metals, as well as other materials, not possible or convenient with other types of heat sources.

Like other solar furnaces, the facility is not really a furnace but an optical system by which solar radiations are concentrated in a small area. The system consists primarily of two mirrors and a target. One mirror, called a 'heliostat,' is flat and movable and is mounted near the ground. It reflects the sun's rays upward to a curved stationary mirror that 'looks' downward, focusing the rays to create a 'hot spot' on a target specimen or crucible about 2 ft. below.

The 'heliostat' is a group of nine back-silvered mirrors mounted on a cradle which is turned by a control system at a rate necessary to track the sun. The mirrors are tilted as a unit to deflect the sunlight toward the curved, polished, aluminium mirror, or paraboloid, 32 ft. above. The target consists of a mounting plate at the end of a retractable framework. Attached to this framework is a shield that can be swung over the target specimen to shut off the heat.

This design provides maximum efficiency in solar heat concentration consistent with the greatest possible advantage in operation and use. Compared with the type of solar furnace in which the entire apparatus moves to follow the sun across the sky, the design permits the target specimen to remain stationary with the specimen or crucible horizontal and facing upwards. Although it requires two mirrors, the second mirror causes only a very small heat loss of 100–150°C.

Temperatures of over 3,000°C. can be obtained by other means, such as actual flame, electrical resistance, arcs or electrical induction. The solar furnace offers, however, many advantages over such methods. The radiant energy obtained does not produce contaminants as do other heat sources. Since the heat is concentrated in a very small area (⅜ in.), the specimen alone is heated.

The heat concentration of the solar furnace permits the quick attainment of high temperatures and subsequent rapid cooling. As the 'hot spot'

Stanford Research Institute's 'solar furnace'

is very sharply defined, it is possible to obtain a sharp temperature gradient, *i.e.* an inch away from the centre of the spot, the temperature drops from over 3,000°C. to over approx. 1,500°C. Also, the movable shield and other types of shutter permit quick shut-off or partial reduction of the applied heat, thereby providing specific heat control.

The new furnace is generally similar in principle to certain furnaces at the Mount Louis Solar Laboratories in the French Pyrenees. The Kennecott Copper Corporation has a somewhat similar facility except that it uses an additional mirror. Specifically, however, the Stanford Research Institute's furnace is the only one of its type in operation in the United States.

Research applications

Work is beginning on a variety of research tasks requiring the kind of high-temperature conditions provided by the new solar furnace. It is particularly useful in the preparation of 'super-pure' materials such as single silicon crystals. Since melting occurs only within the limits of the sharply defined 'hot spot,' and thus on the specimen alone, contamination from a crucible is avoided. This advantage is particularly significant for research on

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The cold extrusion of steel

R. A. P. MORGAN, O.B.E., M.I.Mech.E.

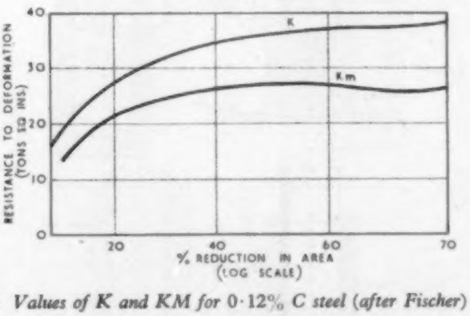
The relatively new field of cold extrusion of steel is surveyed in this article and its techniques discussed with many practical examples. Mr. Morgan is Engineering Director of the War Department. Superintendent, R. O. F. Birtley*

continued from last month

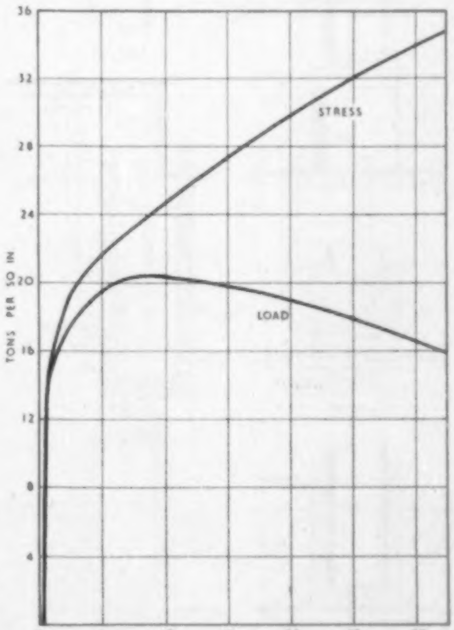
Tool design

THE high loads and stresses imposed upon cold-extrusion tooling is such that unless special measures are taken the tools will plastically deform or fracture. The tool which has to support the whole of the extrusion load by the sheer merit of its strength associated with some ductility is the punch. Its design and manufacture must, therefore, be as perfect as possible. The other tools, namely the die and the die pad, can always be suitably supported to such an extent that the stresses imposed can always be within the elastic range. It must be borne in mind, however, that all tools will elastically strain under load and will, therefore, produce a component of different size to that given by the tools at rest. A steel punch hardened and tempered to about 740 V.P.N. expands under load to a degree dependent upon its diameter. The effect of the extrusion pressure upon such a punch must be considered from two aspects—namely as new and under first trial, and also after having been in use for a considerable run. When first used, the punch, being new, will strain elastically. Evidence seems to point to the fact that diametral strain occurs in two places, namely at the nose and also at a position on the shank from the one-third to one-half length, upwards. Unless some account is made for this in the design of the punch, the effect of the expansion in diameter of the punch from one-third to one-half length position upwards will be to work in concert with the recovery of the die mouth in causing a choke effect at the annulus between the punch and the die mouth which will increase the frictional resistance.

After a punch has been in use for some time

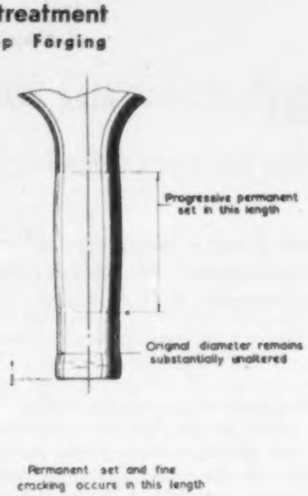


6 *Values of K and KM for* 0·12% *C steel (after Fischer)*



7 *True stress against % reduction in diameter (after Stead)*

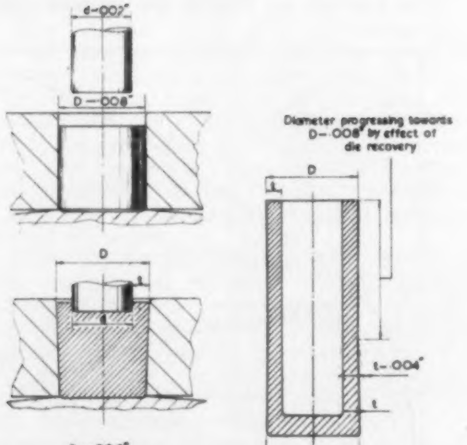
*Article based by the author on his lecture given at the Wolverhampton and Staffordshire College of Technology last March at a two-day symposium on 'Cold flow forming.' The article will be concluded in next month's issue of METAL TREATMENT.



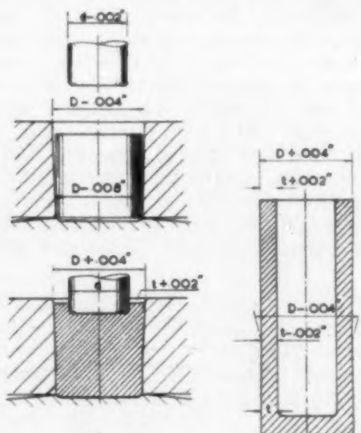
8 *Diagram of punch failure after constant use*

a degree of permanent set occurs and punches removed from service after failure are found to be somewhat as illustrated in fig. 8.

A die is usually made so that it rests on, and is attached to, the die pad. Dies for the cold extrusion of steel are now made of three rings, the inner ring being the die proper and the two outer rings being shrink-rings which impose a compressive hoop stress in the inner ring. By this means the stresses through the die are evened out as far as possible during the extrusion stroke, and the tensile hoop stress on the bore of the die kept down to sustainable limits. In cold extrusion of steel, punch pressures of the order of 120 tons/sq. in. are produced. If this is considered as a fluid pressure, a radial pressure on the inner face of the die of similar order can be expected. In an unsupported die such radial pressure can produce a tensile hoop stress at the die-bore of approximately 136 tons/sq. in. To counteract this it is normal practice to induce into the die bore a compressive hoop stress of about 70 tons/sq. in. by means of the two shrink-rings each imposing 35 tons/sq. in. This would reduce the tensile hoop stress under load to something of the order of 66 tons/sq. in. If a tungsten carbide die were employed, however, no tension



9 *Effect of die expansion linked with uniform expansion of the punch*

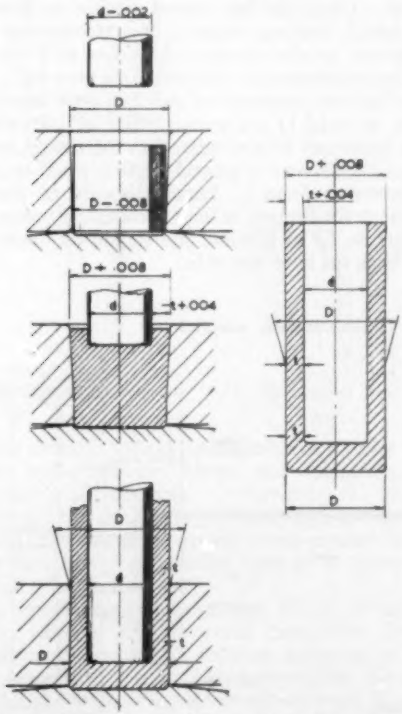


10 *Effect of die expansion linked with uniform expansion of the punch (die tapered 0·004 in. length)*

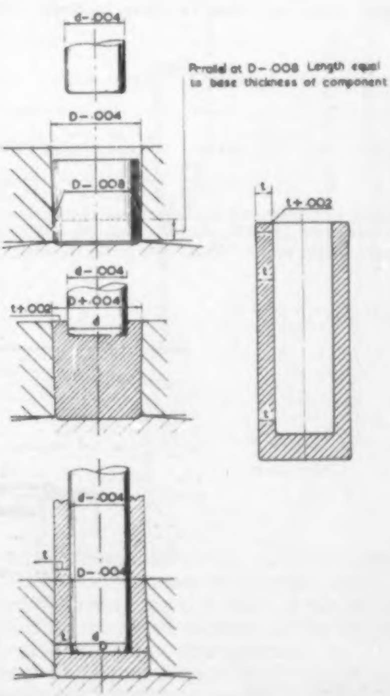
hoop stress under extrusion load can be allowed otherwise the die would crack. In such case a compressive hoop stress equal and opposite to the full tension hoop stress obtained under the extrusion load would have to be applied and shrink-rings of high-quality steel would, therefore, be required.

At first extrusion impact, the mouth of the die is forced to expand and, depending upon the dimensional aspects, the conformity and the method of securement of the die, the expansion at the mouth may, on first impact, be greater than that at the base diameter. However, as extrusion proceeds, the effect is to make the die diameter expand progressively downwards through its depth, whilst the mouth of the die attempts to some degree to recover its original diameter. Whilst this is occurring, the upper portion of the punch shank and the nose of the punch are increased in diameter under load and, as a consequence of the choking effect at the die mouth, frictional resistance to flow upwards becomes very high. The effect of the expansion and recovery of a die is shown in figs. 9, 10, 11 and 12, but is perhaps best illustrated in the production of a high-strength aluminium alloy component which was produced on a Herlan press. This component, see fig. 13, was produced from a tool set-up, as per fig. 14. The interesting facts about this component is that it is waisted in the centre, and it is clear from deduction that at the commencement of extrusion the die expanded at the mouth more than it did at the base. An analysis of the tool set-up shows this to be quite possible. An extrusion lip on the punch 0·024 in. greater in diameter than the punch shank permitted full recovery of the die mouth as extrusion proceeded.

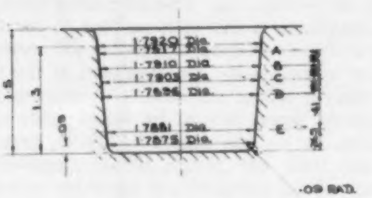
Design of tools for steel extrusion cannot be successful unless the effects upon them of the extrusion load are carefully considered and understood. In considering the punch, the length must be reduced as much as possible in order to reduce undesirable bending effects. The formation of an extrusion lip may well cause the component to be extracted on the punch and this would then require the punch to be lengthened in order to fit a stripper. Whereas, therefore, an extrusion lip on the punch may well reduce the extrusion load, it may not be practicable to allow its use, as the punch length may be increased above the safe margin of



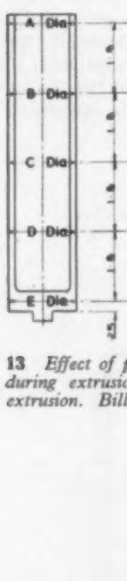
11 *Effect of die expansion linked with uniform expansion of the punch (die tapered* 0·008 *in. length)*



12 *Effect of die expansion linked with preferential punch nose expansion (tapered die)*



COMPONENT CHARACTERISTICS



Diameter Cold	Diameter At 100°C.	Waisting Effect.	Comparison of compt at 100°c against die at rest.
1·7919	1·7942	+·0108	+·0025
1·7848	1·7871	+·0037	−·0039
1·7817	1·7840	+·0006	−·0063
1·7811	1·7834	0	−·0062
1·7899	1·7922	+·0088	+·0041

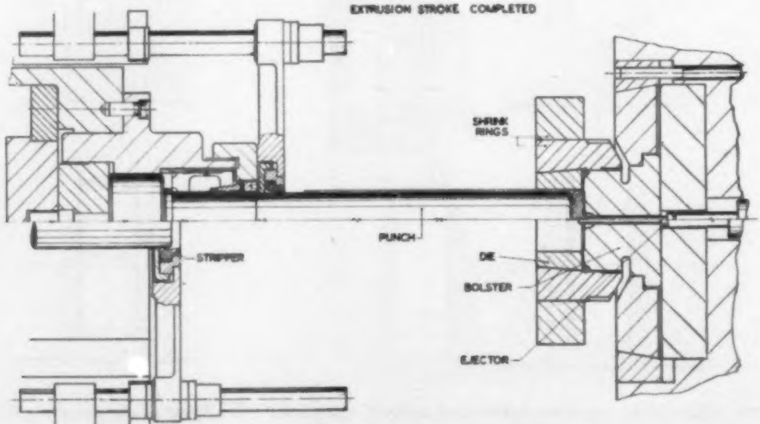
13 *Effect of preferential expansion and contraction of die during extrusion (2·25% MgAl alloy, 50 VPN before extrusion. Billet size 1·785 dia. × 1·305 in. long)*

$2\frac{1}{2}$ to 3 times its diameter. If an extrusion lip is considered essential, its proportions should be somewhat as shown by fig. 15. Bearing in mind that the punch must seat against a pressure pad, the designer may effect a design as in fig. 16. This, however, is not good from the heat-treatment point of view as well as being expensive for replacement, and fig. 17 represents a better design in regard to these aspects.

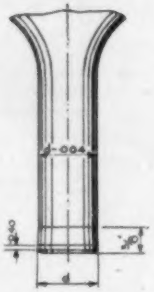
Except for the fact that the greater the number of parts in any set-up the more risk there is of dimensional and flatness inaccuracies, a build-up of pads as shown by fig. 18, bringing the eventual load evenly over the press platen, is preferred to solid tooling both from the economy and heat-treatment points of view. The flatness and parallelism of the pads is particularly important and all such pads should be surface ground.

The punch nose surface is preferred to be flat, finishing with a small radius to form the side wall of the punch. The size of the radius is important as it has some effect upon the pressures required for extrusion. The flat surface of the punch serves to maintain the phosphate and the lubrication without fracture and this keeps the punch loading down. Once the lubrication surface is broken, the punch loadings required rise steeply and the graphical results showing the effect of a conical surface on the punch illustrates this very well.

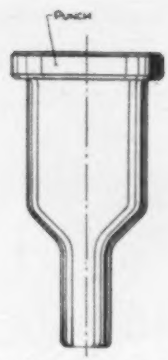
In the heat treatment of punches, great attention must be paid to the surface prior to quenching. It is important to remember that there must be no stress raisers and a ground surface prior to heat treatment is desired. Punches should be packed whilst in the furnace to limit preferential heating or cooling as far as possible and should be quenched out from the nose upwards.



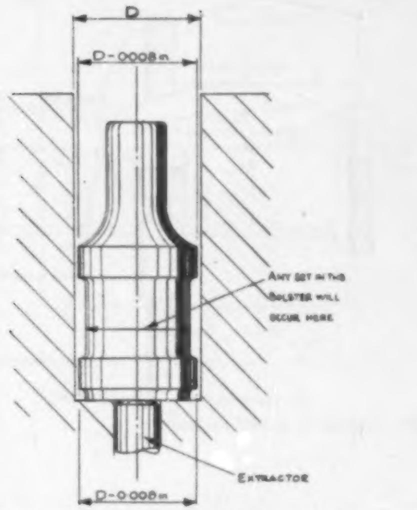
14 *Tool set-up as used for production of high-strength aluminium alloy component*



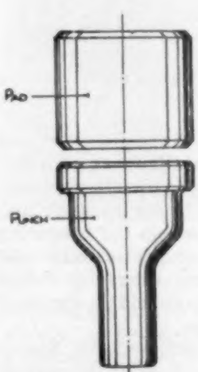
15 *Extrusion lip proportions*



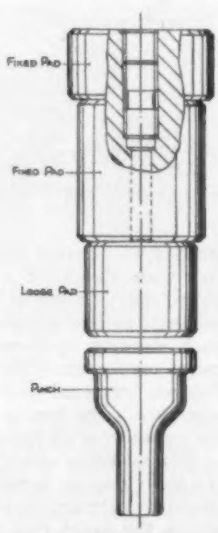
16 *Design not good for heat treatment*



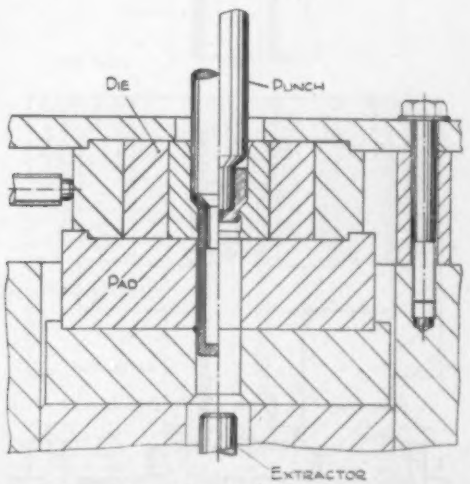
19 *Bolster designed to expand under load in a preferred position*



17 *Better design from heat-treatment standpoint*



18 *Build-up of pads*

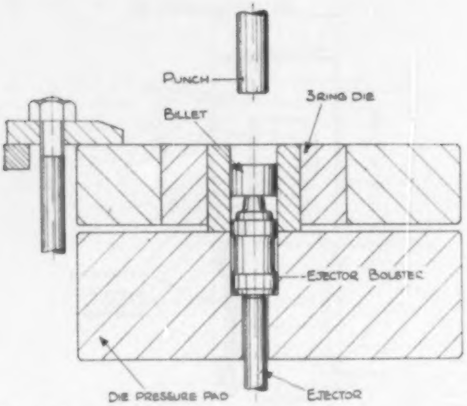


20 *Typical Hooker extrusion*

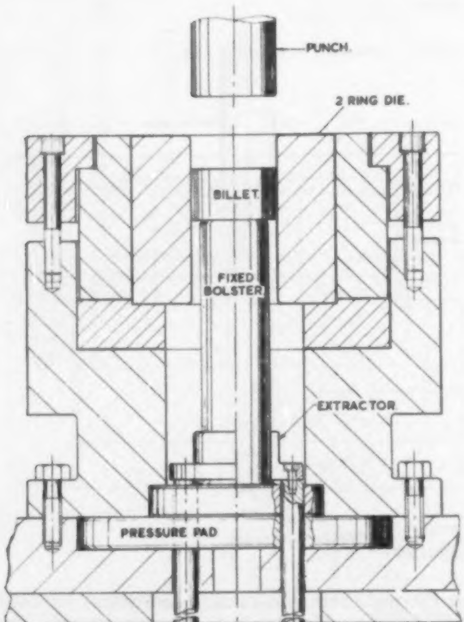
The bolster should be treated in exactly the same way from the design and manufacturing standpoint as the punch. If, however, the bolster is to be used also as an extractor then the design should be such as to enable it to expand under load in a preferred position such as is shown in fig. 19.

In considering die design, figs. 20, 21, 22 and 23 show different set-ups which have been found satisfactory. In fig. 23, troubles occurring in the maintenance of a tight joint between the die and the die pad were overcome by allowing the die pad to set down elastically under load in the position shown. When this die is securely screwed on to the die pad there is a turning movement imposed upon the die which tends to open the mouth. In a properly stressed die this effect would be small; nevertheless, the mouth diameter of the die should be checked before and after assembly.

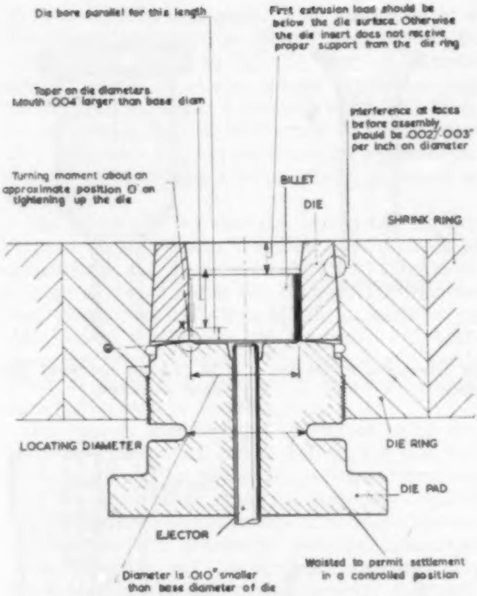
The most important part in die design lies in the accurate stressing and proportioning of the die proper and the shrink-rings. Experience has shown that all shrink-rings and the die proper have an



21 *Typical backward extrusion*



22 *Typical forward extrusion*



23 *Die design for expansion and settlement under load*

internal to external diameter ratio of 1 : 1·6 and that there should be a shrink allowance of 0·003 in./in. dia. To enable this, it is usually necessary to warm up the shrink rings before assembly and such warming up should be within the tempering range employed. The die proper is fully heat treated, tempered and ground externally before assembly. The bore is finally ground to size after the assembly of all shrink rings.

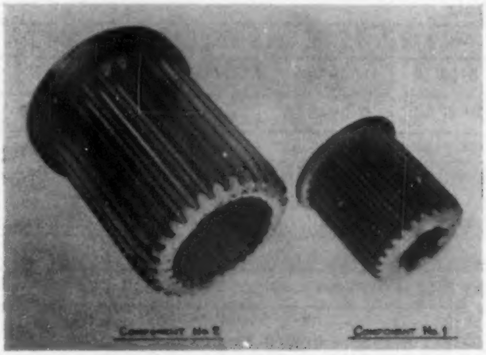
The assembly order is as follows: (1) Shrink the inner ring on to the die; (2) grind the outside diameter accurately to size after shrinking; and (3) shrink the outer ring on to the assembled inner ring and die.

The shrink-ring material usually employed is similar to En 25, namely C, 0·27–0·35% ; Ni, 2·3–2·8; Cr, 0·5–0·8; Mo, 0·4–0·7%.

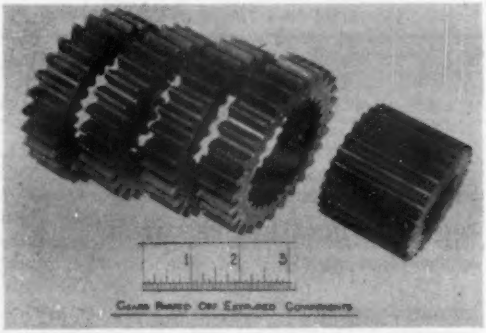
After finish turning, all shrink-rings are heat treated to about 80 tons/sq. in., tempered at about 250°C. The shrink-rings are then warmed up to 350°C. to expand them sufficiently for engagement with the mating rings and allowed to cool in position. Parallel bores are preferred to tapered bores, being simpler to manufacture and to measure.

Cold extrusion of gears

Realizing the economic factors mitigating against the manufacture of simple tubular components by cold extrusion, attention has been paid to the



24 *Cold-extruded gear blanks*



25 *Gears produced from extruded components*

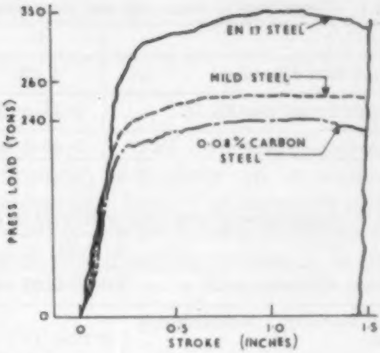
manufacture of more difficult objects such as gear wheels. A process was developed on the basis of a consideration of the volume of material required in order to produce an extrusion using the smallest power, and certain success has been experienced in making gear blanks in the three steels shown in table 1. These components are illustrated in fig. 24 and the gears produced therefrom in fig. 25.

All finished components were stress relieved at 400°C. for 1 h. after which the physical characteristics observed were as in table 2. Stroke/load curves are as in fig. 26. Section and tooth hardness figures were as shown in figs. 27 and 28.

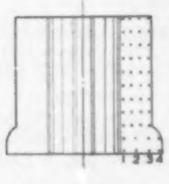
Two types of gear blanks were made and it will be noted that the internal splines and gear teeth were formed at the same time. Dimensional characteristics are as follows:

Component No. 1

EXTERNAL: Outer dia. 2·655 + 0·002 in. Root dia. 2·338 in. Pressure angle 25 deg. –27 teeth.



26 *Stroke/load curves for gears*

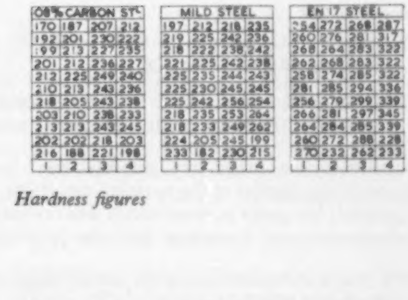


·08% CARBON STᴸ			
170	187	207	212
192	201	230	222
199	213	227	235
201	212	236	227
212	225	249	240
210	213	243	236
218	205	243	238
203	210	238	233
213	213	243	245
202	202	218	203
216	188	221	198
1	2	3	4

MILD STEEL			
197	212	218	235
219	225	242	236
218	222	238	242
221	225	242	238
225	235	244	243
225	230	245	245
225	242	256	254
218	235	253	264
218	233	249	262
224	205	245	199
233	182	230	215
1	2	3	4

EN 17 STEEL			
254	272	268	287
260	276	281	317
268	264	283	322
262	268	283	322
258	274	285	322
281	285	294	336
256	279	299	339
266	281	297	345
264	284	285	339
260	272	288	228
270	232	262	233
1	2	3	4

27 *Hardness figures*



28 *Hardness figures*

TABLE 1 *Compositions of three steels used for making gear blanks*

	C	Mn	Si	Ni	Cr	Mo
0·08% C fully-killed steel	0·08	0·12	0·07	—	—	—
Commercial mild steel En 3A ..	0·15–0·25	0·40–0·90	0·05–0·35	—	—	—
Manganese-molybdenum steel En 17	0·30–0·40	1·30–1·80	0·10–0·40	—	—	0·35–0·55

TABLE 3 *Compositions of some other steels tried for gear making*

	C	Mn	Si	Ni	Cr	Mo
En 18 1% chromium steel ..	0·35–0·45	0·60–0·95	0·10–0·35	—	0·85–1·15	—
En 361 low-alloy case-hardening steel	0·13–0·17	0·70–1·00	0·35	0·40–0·70	0·55–0·80	0·08–0·15
SAE 8615 low-alloy case-hardening steel	0·13–0·18	0·70–0·90	0·20–0·35	0·40–0·70	0·40–0·60	0·15–0·25
S 82 4¼ Ni-Cr-Mo case-hardening steel	0·12–0·18	0·50	0·10–0·35	4·00–4·50	1·00–1·40	0·15–0·35
S 106 nitriding steel	0·20–0·25	0·10–0·35	0·40–0·65	0·30 max.	2·90–3·50	0·40–0·70
S 107 3% Ni-Cr-Mo case-hardening steel	0·18	0·10–0·35	0·30–0·60	3·00–3·75	0·60–1·10	0·10–0·25

INTERNAL: Six splines width 0·342 in., depth 0·105 in. Bore dia. 1·19–0·0005 in.

Component No. 2

EXTERNAL: Outer dia. 3·845 + 0·005 in. Pressure angle 20 deg.–28 teeth.

INTERNAL: Internal dia. 2·370 in. Root dia. 2·662 in. Pressure angle 25 deg.–27 teeth.

Recently various other steel qualities have been put through the smaller of the two tool sets (table 3).

In general, the gears in these steels were required for subsequent heat treatment and the properties

TABLE 2 *Physical characteristics of the finished components made from steels in table* 1

	U.T.S. T.S.I.	Y.S. T.S.I.	Elong. $4\sqrt{A}$ %	R. of A %	Izod ft./lb.
0·08%	42·2	41·8	14	54	8
M.S.	52·2	50·7	11½	44	13·3
En 17	61·8	57·2	12	47	14·8

attained by cold working are of no consequence. The extrusion loads for En 18 and En 361 were in the same order as for the En 17 steel illustrated, whilst those for the remaining three were in the region of 375–400 tons.

to be continued

Solar furnace for high-temperature research

concluded from page 450

transistor materials where extreme purity is important.

Measurement of thermal characteristics and conductivity of materials at high temperatures is easier with the solar furnace. To study the effect of thermal shock on ceramic material, for example, a chilled specimen can be placed on the target plate under the movable shield. By suddenly withdrawing the protecting shield, the ceramic is exposed instantly to the furnace temperature.

Other studies include the effect of high temperatures on electrical conductivity of metals, on materials under controlled gaseous or other environmental conditions, and on the ignition tendencies and characteristics of solid propellants and other explosive agents. The rates of material reaction under quick and great changes in temperature can be studied and the temperature and environmental characteristics of the upper atmosphere simulated.

Design and construction of the new solar furnace represents a co-operative effort of the Institute directed by Dr. Nevin K. Hiester, manager of chemical engineering research, with major assistance by chemical engineers Raymond K. Cohen and Rodney B. Beyer.

metal treatment
and Drop Forging

The creep of metals

P. FELTHAM, D.Sc., F.Inst.P.

Recent advances in experimental techniques, in particular the development of etch-pit and electron transmission micrography, have led to appreciable advances in the understanding of the role of dislocations in the plasticity of metals. In the present paper the theoretical basis of the creep of metals is reassessed in the light of these observations, and some of the major problems requiring further study are discussed*

THERE NOW APPEARS to be general agreement that the deformation of metals occurring under constant stress, which may be quite rapid at temperatures comparatively high with respect to the melting point, arises from a balance between work-hardening and recovery. The increasing resistance of the metal to further deformation therefore tends to be counteracted by softening, so that creep can continue to take place. It is possible to separate these two opposing tendencies and to study them more or less in isolation by taking advantage of the fact that at sufficiently low temperatures work-hardening can still occur while recovery is virtually absent.

Such an experiment could be carried out as illustrated in fig. 1. Here the metal is first extended plastically at a comparatively low temperature up to a certain strain indicated by the point 'A.' The metal, for example iron, is then removed from the tensile testing machine and heated for some time at a high temperature, say a few minutes at 800°C. in our case. After cooling its flow stress will be the same, or nearly the same, as initially, and on applying the same load as before it will be strained to point 'B.' Hence, by allowing recovery to intervene, we have been able to double the strain without increasing the load.

Procedures of this type are, of course, well known from metal working operations, such as cold-rolling or wire drawing, when the metal has to be annealed between passes, but the bearing of this unity between work-hardening and softening by heat treatment is not always fully appreciated in connection with creep, in which both processes occur simultaneously. However, the interpretation of creep as a 'struggle' between hardening, which induces structural changes in the material, and recovery, which opposes and tends to reverse the changes due to strain, is basic to any theory of creep in metals. It is also rather general, and is probably valid for all work-hardening solids.

Formally, one can express the observation that an increment of the flow stress $d\sigma$ may comprise contribution due to increments of the strain ε, the temperature T (which may affect the elastic constants) and the time of isothermal annealing by writing:

$$d\sigma = \frac{\partial\sigma}{\partial\varepsilon} d\varepsilon + \frac{\partial\sigma}{\partial T} dT + \frac{\partial\sigma}{\partial t} dt \quad \ldots\ldots\ldots\ldots(1)$$

This equation may be integrable with certain specific boundary conditions, but we shall not concern ourselves with this at the moment. In conventional creep tests carried out at constant stress and temperature, equation (1) simplifies considerably, for $d\sigma$ and dT are then zero, yielding:

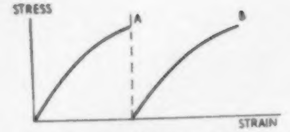
$$d\varepsilon/dt = -(1/\chi)\ \partial\sigma/\partial t \quad \ldots\ldots\ldots\ldots\ldots\ldots(2)$$

which shows particularly clearly the unity between the rate of work-hardening $\chi = \partial\sigma/\partial\varepsilon$ and the rate of recovery $\partial\sigma/\partial t$ already referred to.

At the next, higher, level of enquiry it thus becomes necessary to investigate the modes of work-hardening and recovery specific to metals. As these phenomena are not independent of one another the problem of correlating them also arises. Although the principal features of both processes, each considered in isolation as in the example in fig. 1, are now beginning to be reasonably well understood, it has not as yet proved possible to integrate them into a quantitative theory of the cybernetics of creep.

At this juncture, one may thus be inclined to enquire what the major unsolved problems are and how to attempt to solve them most rationally. How-

* The author, of the Department of Metallurgy, The University, Leeds, has based this article on his contribution to a colloquium held at the Academy of Sciences in Budapest on August 4, 1961.



1 *Doubling of the strain facilitated by an intermediate anneal at 'A'*

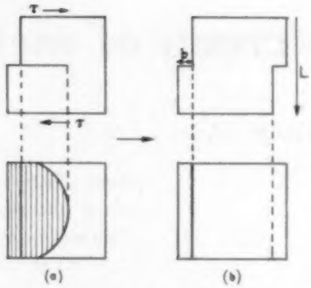
ever, obtaining an answer to this is not a straightforward exercise in formal logic which could be sorted out by an electronic computer. Some of the tasks will emerge in the course of this discussion but, like with any other branch of research or social activity in general, the question as to the best approach and the fastest advance cannot be examined adequately without tracing, in some detail, the historical development of the subject in its concrete social setting.

Such an analysis[1] may be rather fruitful, because thinking about research in this broad context leads one to appreciate the present growing points of the science, the effect of the allocation of funds to research on the rate of growth of any given branch of investigation, and so on. Sometimes such an analysis also helps one to spot questions which today exist only 'in embryo,' but which will eventually present themselves as fully matured problems for solution when advancing technology encounters them. However, it would not be possible to do justice to such a study within the framework of our present aims, and the object of raising it was mainly to stress its relevance; in relation to any purposive activity such an analysis could serve as a better councillor as to what to do and how to do it than the intuitive empiricism by which similar questions are often decided.

Here, then, I want to limit the discussion mainly to the nature of work-hardening and recovery, and their probable manner of interaction in creep. Apart from passing references to alloys I shall deal foremostly with pure metals, for although much useful knowledge has been accumulated about structures desirable in alloys having high creep strength, generally by empirical or semi-empirical means, and a deeper insight is rapidly developing as a result of the application of advanced techniques in research, the interpretation of many features is still rather controversial; in attempting to discuss them we might be led too far into speculation.

Dislocation kinetics

We shall take as our starting point the now well-established fact that plastic deformation of metals is facilitated by dislocations, and before considering the specific problem of plastic deformation in creep at elevated temperatures in more



2 *The partial and complete passage of an edge dislocation of Burgers vector b through a crystal cube of edge L*

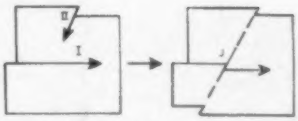
detail it will be useful to examine briefly some of the characteristics of dislocations which will be particularly relevant in dealing with creep. Fairly extensive accounts of the basic properties of dislocations exist[2, 3] and we shall review only those which have a direct bearing on our present problem.

First, we consider the passage of a straight edge dislocation through a crystal cube of length of side L, as shown in fig. 2*a*. The cube may be taken to represent a grain in a polycrystalline metal. The shaded area denotes the zone which has already slipped; the dislocation represents the boundary between the slipped and unslipped parts of the glide plane. When the dislocation has passed through the crystal, as shown in fig. 2*b*, the shear strain will be $\mathbf{b}/L$, where $\mathbf{b}$ is the Burgers vector of the dislocation. If more than one dislocation move simultaneously through the crystal, then with ρ taken as the density of similar *moving* edge dislocations per unit area perpendicular to the glide plane, the strain induced in the average time t_L required by a dislocation to traverse the crystal will be given by $(L^2\rho)(\mathbf{b}/L)$, and the shear rate will therefore be $L\rho\mathbf{b}/t_L$. If we denote the mean velocity of the dislocations by $\bar{v}$, where $\bar{v} = L/t_L$, then the creep rate in shear becomes:

$$d\gamma/dt = \rho\mathbf{b}\bar{v} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots (3)$$

This simple expression is useful, first, because in dealing with a specific, even though not quite realistic, model it is already somewhat less general than the basically equivalent relation given by equation (2), and second, because in throwing into relief the two important variables determining the creep rate of the model it enables us to decide which of the properties of dislocations will be particularly relevant to our discussion of the creep in real metals. It should be emphasized that ρ refers only to dislocations participating in creep, the total density of dislocations in a real crystal may be higher.

The real crystal will differ from the idealized one

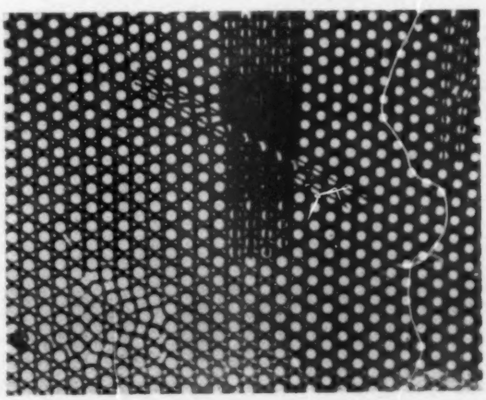


3 *Formation of a jog 'J' by intersection of two dislocations*

in several other respects, *e.g.* dislocations will propagate on a number of intersecting systems of glide planes and intereact with one another to give tangles or more regular networks. In general more than one slip system will be operative at any given time in the process of deformation; in crystals representing grains of polycrystalline aggregates several slip systems must, in fact, operate simultaneously if the metal is to remain continuous and is not to break to pieces.[4]

The interaction of dislocations in the course of plastic deformation of real crystals results in structural changes which affect both ρ and $\bar{v}$. An example of slip on two intersecting slip planes is shown in fig. 3; the formation of steps, known as 'jogs,' in one of the dislocations due to the passage through it of another is illustrated. Although the jog is essentially a small piece of edge dislocation it will not move as readily as a longer edge dislocation because (*a*) if it moves at right angles to the plane of the paper, as the dislocation loop containing it expands under an increasing shear stress, the derangement of the crystal lattice at the corners of the Z-shaped imperfection of which it is part will exercise a temperature dependent frictional force on the dislocation containing the jog,[5] while (*b*) if it moves in the plane of the paper, say from left to right, thus causing the lower part of the dislocation to 'climb' up to the upper part, it must either generate vacancies or await the arrival of

4 *A two-dimensional edge dislocation in a soap-bubble raft*



vacancies through self-diffusion from the body of the crystal.†

How vacancies can induce jog migration, and hence dislocation climb, may be seen more clearly from fig. 4, which shows a two-dimensional edge dislocation in a soap-bubble raft. The dislocation can conveniently be regarded as arising from the insertion of an extra plane of atoms from above into the crystal. To make it climb up a line of atoms would have to be removed from the bottom edge of the plane; this can be done if the atoms at the edge can escape by changing place with vacancies situated adjacent to them. Energetically the displacement of an existing jog along the edge dislocation is therefore equivalent to vacancy migration.

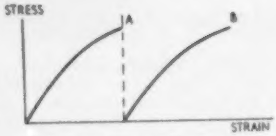
Dislocation displacements of this type, in which point defects are released or absorbed, are known as 'non-conservative,' to distinguish them from the conservative displacements, *e.g.* that of a jog perpendicular to the plane of the paper (fig. 3), when point defects are not formed. As a result of the drag exercised by jogs the migration velocity of dislocations will depend not only on stress but also on temperature.

Edge dislocations cannot then leave their slip planes and move to parallel adjacent ones, except non-conservatively, by climb. Screw dislocations, however, can move to adjacent slip planes conservatively, and this ease of transfer from one slip plane to another probably has as a consequence the well-known weavy appearance of slip bands observable, for example, in alpha-iron.

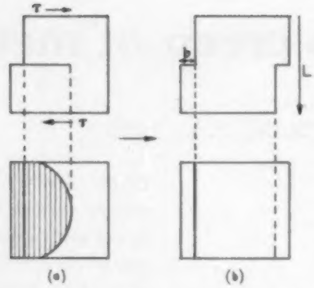
In face-centred cubic metals such 'crosslip' by screw dislocations is more difficult, because the glide dislocations split up into adjacent half-dislocations separated by a thin ribbon of stacking-fault, and to get such an extended dislocation to crosslip is somewhat like trying to push a fairly long cylindrical coil spring through a sharp bend in a narrow pipe; the two half-dislocations, just like the two ends of the spring, would first have to be forced together, and high stresses would be required. However, even in face-centred cubic metals crosslip could be induced by the cooperation of screw dislocations of opposite signs, for example as shown in fig. 5.

The diagram shows two successive stages in the approach of two mutually attracting screw dislocations lying on intersecting slip planes, resulting in their eventual annihilation. Interactions of this kind, resulting in the loss of dislocations, or segments of dislocations, induce recovery, as we shall see below. The mechanism illustrated in fig. 5 does not require thermal activation, except in so

†As the jog is inflexible it may also have to overcome a frictional force akin to the Peierls force.[3]



1 *Doubling of the strain facilitated by an intermediate anneal at 'A'*



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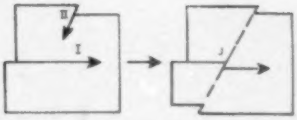

3 *Formation of a jog 'J' by intersection of two dislocations*

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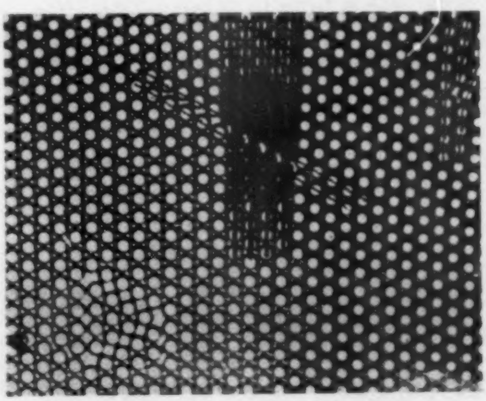
The interaction of dislocations in the course of plastic deformation of real crystals results in structural changes which affect both ρ and $\bar{v}$. An example of slip on two intersecting slip planes is shown in fig. 3; the formation of steps, known as 'jogs,' in one of the dislocations due to the passage through it of another is illustrated. Although the jog is essentially a small piece of edge dislocation it will not move as readily as a longer edge dislocation because (*a*) if it moves at right angles to the plane of the paper, as the dislocation loop containing it expands under an increasing shear stress, the derangement of the crystal lattice at the corners of the Z-shaped imperfection of which it is part will exercise a temperature dependent frictional force on the dislocation containing the jog,[5] while (*b*) if it moves in the plane of the paper, say from left to right, thus causing the lower part of the dislocation to 'climb' up to the upper part, it must either generate vacancies or await the arrival of

4 *A two-dimensional edge dislocation in a soap-bubble raft*



vacancies through self-diffusion from the body of the crystal.†

How vacancies can induce jog migration, and hence dislocation climb, may be seen more clearly from fig. 4, which shows a two-dimensional edge dislocation in a soap-bubble raft. The dislocation can conveniently be regarded as arising from the insertion of an extra plane of atoms from above into the crystal. To make it climb up a line of atoms would have to be removed from the bottom edge of the plane; this can be done if the atoms at the edge can escape by changing place with vacancies situated adjacent to them. Energetically the displacement of an existing jog along the edge dislocation is therefore equivalent to vacancy migration.

Dislocation displacements of this type, in which point defects are released or absorbed, are known as 'non-conservative,' to distinguish them from the conservative displacements, *e.g.* that of a jog perpendicular to the plane of the paper (fig. 3), when point defects are not formed. As a result of the drag exercised by jogs the migration velocity of dislocations will depend not only on stress but also on temperature.

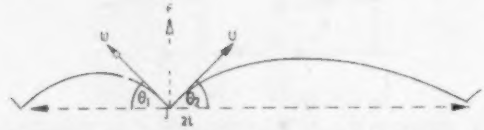
Edge dislocations cannot then leave their slip planes and move to parallel adjacent ones, except non-conservatively, by climb. Screw dislocations, however, can move to adjacent slip planes conservatively, and this ease of transfer from one slip plane to another probably has as a consequence the well-known weavy appearance of slip bands observable, for example, in alpha-iron.

In face-centred cubic metals such 'crosslip' by screw dislocations is more difficult, because the glide dislocations split up into adjacent half-dislocations separated by a thin ribbon of stacking-fault, and to get such an extended dislocation to crosslip is somewhat like trying to push a fairly long cylindrical coil spring through a sharp bend in a narrow pipe; the two half-dislocations, just like the two ends of the spring, would first have to be forced together, and high stresses would be required. However, even in face-centred cubic metals crosslip could be induced by the cooperation of screw dislocations of opposite signs, for example as shown in fig. 5.

The diagram shows two successive stages in the approach of two mutually attracting screw dislocations lying on intersecting slip planes, resulting in their eventual annihilation. Interactions of this kind, resulting in the loss of dislocations, or segments of dislocations, induce recovery, as we shall see below. The mechanism illustrated in fig. 5 does not require thermal activation, except in so

†As the jog is inflexible it may also have to overcome a frictional force akin to the Peierls force.[3]

5 *The approach of two screw dislocations leading to their mutual annihilation*



6 *Forces due to the dislocation line tension U acting on a jog J*

far as the interacting dislocations are jogged; in that case they would experience a strongly temperature-dependent frictional drag, as jogs in dislocations having a large screw component can generally move with the dislocation only non-conservatively,[2] *i.e.* by generating vacancies or by migrating to vacancies which have diffused to positions adjacent to them.

Activation energies

Analyses of the kinetics of dislocations containing jogs, with a more detailed discussion of the various effects of jogs on their movement, have been made by Seeger[6] and Feltham and Meakin[7] in greater detail than is possible here. In all cases the considerations are based on models of the type shown in fig. 6, which shows a dislocation, lying in the plane of the paper, subjected to a shear stress τ. The dislocation, which behaves somewhat like an elastic string with a constant line tension U, exerts a force on the jogs impeding its movement. If, for example, the Burgers vector of the dislocation is along the line F, so that the dislocation is almost purely 'edge,' then the force along F, *i.e.* in the direction of the easy, conservative, movement of the jog, where the energy barrier is comparatively low, will be U ($\sin \theta_1 + \sin \theta_2$), while the force perpendicular to F, tending to move the jog non-conservatively against a high energy barrier will be much smaller, amounting only to U ($\cos \theta_1 - \cos \theta_2$). Clearly, in this case the jog will move conservatively.

If the dislocation were almost purely 'screw' then the behaviour of the jog is less easily ascertained. For example, if θ_1 and θ_2 are almost equal, the jog will move non-conservatively along F, otherwise it may run off sideways until it is either annihilated on encountering a jog of opposite sign or stopped on reaching parts of the dislocation which are of mixed edge/screw type.

If a dislocation is imagined as a circular loop then it is clear that pure edge and pure screw components will exist only over very small lengths of arc at points where the circle is intersected by diameters perpendicular and parallel to the Burgers vector of the dislocation respectively. As the possibility of easy conservative movement exists in both cases it is to be expected that jogs will not concentrate in these sections; more likely positions for the accumulation of jogs will be those parts of the circle where conservative and non-conservative movements of jogs can occur with equal ease, *i.e.* on parts which are of the mixed edge/screw type.

The displacement of jogs in such regions would necessitate the formation or absorption of vacancies, and the dislocation would simultaneously glide and climb. Evidence of the type of dislocation movement involving jogs appears to be apparent at 'A' in the two consecutive electron micrographs of a thin film of copper (fig. 7), which had been prepared from a polycrystalline specimen after creep at 500°C., and which underwent further deformation due to stresses induced by contaminating surface films formed on the specimens during observation in the microscope.

A simple analysis based on classical dislocation theory[8] shows that in a configuration such as is shown in fig. 6 the energy barrier to jog migration, and hence to dislocation movement, decreases linearly with the shear stress τ acting on the dislocation in the slip plane along its Burgers vector, and one may write:

$$E(\tau) = H - \tau \mathbf{b}^2 l \qquad \ldots\ldots\ldots\ldots\ldots\ldots\ldots (4)$$

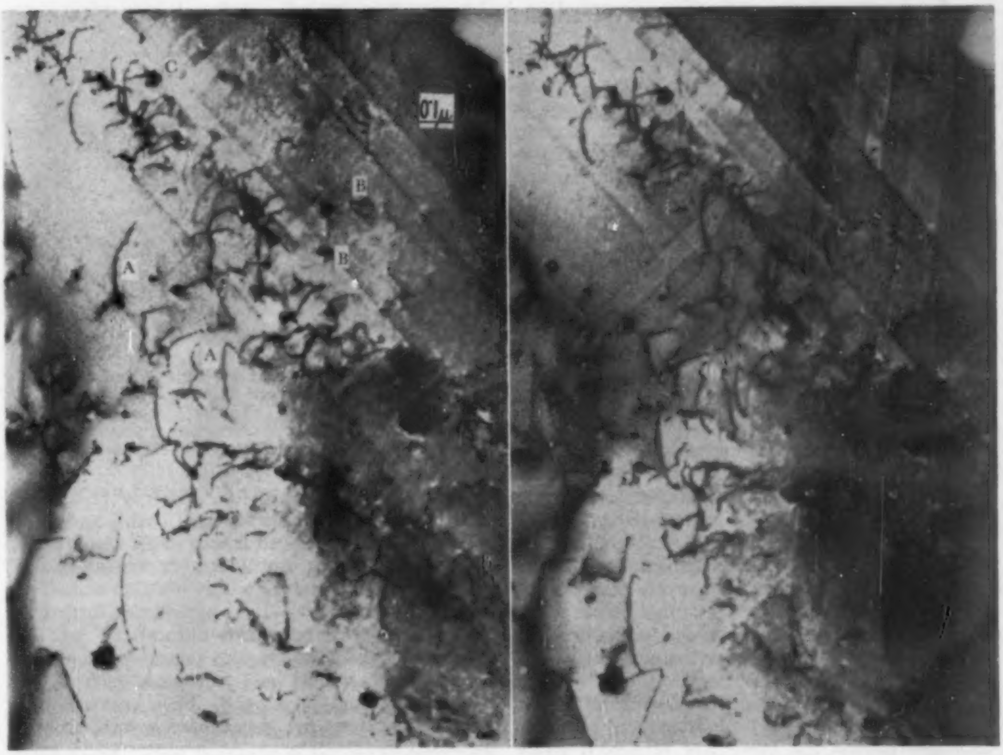
where H is the height of the barrier in the absence of a stress, and l the spacing between jogs as indicated in fig. 6. One can represent equation (4) formally also by writing $\tau_y = H/l\mathbf{b}^2$, so that

$$E(\sigma) = H[1-(\sigma/\sigma_y)] = H[1-(\tau/\tau_y)] \qquad \ldots\ldots\ldots\ldots (5)$$

where the τ's and σ's refer to shear and tensile stresses respectively; the latter case being useful in considering creep in polycrystals. The significance of σ_y may be seen by referring to the velocity of the jog, which may be expressed by the usual rate equation:

$$v_j = \nu \mathbf{b} \exp \{-H[1-(\sigma/\sigma_y)]/\boldsymbol{k}T\} \qquad \ldots\ldots\ldots\ldots (6)$$

in which ν is an atomic frequency of the order of 10^{11}/sec., $\boldsymbol{k}$ Boltzmann's constant, and T the temperature expressed in °K. For if we imagined the specimen containing the dislocation under consideration to be quenched rapidly under load from T to close to 0°K. then, if we wished to maintain the creep rate due to v_j at T, we would have to increase σ to σ_y or, more precisely to



7 *Dislocation movement in a thin film of copper prepared from a polycrystalline specimen after 10% strain in creep at 500°C. under a tensile stress of 475 kg./cm.² and an equilibrium strain rate of 2·7 × 10⁻⁵/sec. Interval between exposures approximately ½ min. Migration at 'A,' shrinking loops at 'B,' cluster of small stable loops at 'C.' Cooled under stress*

$\sigma_y(G_0/G)$, where G_0 and G are the shear moduli at T and 0°K respectively. Thus σ_y may be regarded as the stress required to move the jog with velocity v_j at very low temperatures; in so far as τ_y depends upon the substructure dimensions and geometry (equations 4 and 5), so will then also σ_y.

We now have to consider the magnitude and significance of H. To this end it is useful to employ equation (6), rewriting it in the form

$$\sigma = \sigma_y\,[1 - m(\boldsymbol{k}T/H)] \quad \ldots\ldots\ldots\ldots\ldots\ (7)$$

where

$$m = \ln\,(\nu\mathbf{b}/v_j).$$

If one makes the reasonable assumption that if creep is readily observable then the jogs must run along a dislocation segment of the order of one micron in length in a few seconds, *i.e.* taking v_j equal to about 10^{-5} cm./sec., then with $\mathbf{b} = 3 \times 10^{-8}$ cm. one obtains

$$m \simeq 20 \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ (8)$$

and this value would vary by only approximately ± 10% if v_j were assumed either larger or smaller by a factor of 10 than the value used in deriving equation (8).

On using as an example $H = 1\cdot3$ eV, as observed by Feltham and Meakin[7] in creep of 99·99% oxygen-free high conductivity copper, one finds that the bracketed term in equation (7) is zero at about 480°C., implying that a 1·3 eV barrier should cease to provide creep strength at that temperature. In fact, Feltham and Meakin find that at about 500°C. the activation energy begins to increase rapidly, eventually reaching a value equal to the activation energy of self-diffusion (2·1 eV), close to 700°C. Sherby, Lytton and Dorn[9] find similarly that H in pure aluminium ranges from about 0·1 to 1·0 times the activation energy of self-diffusion as the creep temperature is raised from 77 to 880°K. It appears therefore that more than one value of H, and hence more

than one mechanism of creep, can occur and must be considered.

The lowest barriers to the movement of dislocations would be operative at the lowest temperatures. Glen,[17] for example, observed creep in cadmium crystals at 1—5°K. The extent of creep was very limited, just as in the case of the logarithmic creep observed by Feltham[8] in copper and alpha-brasses at 77—300°K. The former type of creep could be due to the conservative movement of jogs in screw dislocations to points on the dislocations where they are held up by the increasing edge-component of the dislocation; the latter form of creep appears to be due to a similar restricted movement of edge dislocations.[8] Higher activation energies, close in magnitude to the energies of formation of vacancies, migration of vacancies, and self-diffusion could be interpreted as arising from the movement of jogs in fast moving dislocations by Seeger's mechanism, and by the migration of jogs in dislocations having predominantly edge or screw character respectively.

The movement of a jog in a mixed dislocation induces slip and climb simultaneously; the climb could take place by the displacement of the jog facilitated by the uptake of vacancies adsorbed on the dislocation.

Such a process could be regarded as consisting of climb in a vacancy supersaturation, and the observation of activation energies close to that of vacancy migration in the creep of copper has, in fact, been explained in this manner by Feltham and Meakin,[7] who also discuss the movement of jogged screw dislocations associated with an activation energy equal to self-diffusion, using a model as shown in fig. 6. A similar mechanism appears to occur also in the creep of alpha-brasses,[10] but the jog kinetics seem to be rather more complex there than in copper.

The fact that a single activation energy may not suffice to account for the characteristics of creep of pure metals even at comparatively high temperatures must therefore be taken into account in the theoretical treatment of creep. If two or more interdependent rate processes occur simultaneously, *e.g.* one leading to intragranular and one to intergranular recovery, then whichever of the two is the slower one at any given temperature will control the creep rate; at another temperature, however, the roles may become reversed.[7, 10]

Work-hardening and recovery

In the discussion of the effects of the mutual intersection of dislocations on their mobility under stress we found that it led to the emergence of a temperature dependent frictional drag. We also referred to one form of dislocation/dislocation interaction which led to recovery. If, according to the

8 *Dislocation interactions. 'A' shows two L-shaped nodes, possibly due to interaction of an edge and screw dislocation of opposite Burgers vectors; 'B' triple nodes. Some of the dark clusters were observed to contain small loops. Specimen as in fig. 7, but strain rate somewhat less*

suggestion previously made, recovery results in, and from, a loss of dislocations, then one should also expect that work-hardening would result in the converse, *i.e.* in an increase of the density of dislocations in the metal.

Before considering this hypothesis further it should be remarked that dislocations will in general be stabilized by jogs or/and by comparatively strong nodes‡ and junctions, such as are shown at 'A' and 'B' in fig. 8. This is not a restrictive assumption, for dislocations which are not impeded in this or a similar manner would move under extremely small stresses, and would therefore either glide out of the crystal or, on encountering existing networks, join them or interact with them destructively. Thus the resistance to deformation must be regarded as deriving from dislocations which, as a result of interaction, have become 'collectively organized' in comparatively stable configurations, such as networks locked by dislocations of the Cottrell-Lomer type,[2] polygon walls or other sub-boundaries.

Now, a correlation between progressive work-hardening and increasing parcellation of the crystal into comparatively strain-free 'blocks'[11] bounded by dislocation sub-boundaries has been suspected for many years, and recent researches utilizing micro-focus X-ray tubes, etch-pit methods and electron microscopy have confirmed the basic

‡The empty space between the two L-shaped junctions at 'A' may be due to annihilation of part of two dislocations on intersection, *e.g.* where a screw dislocation on a $(11\bar{1})$ plane with Burgers vector $[1\bar{1}0]$ intersects an edge dislocation on a (111) plane with opposite Burgers vector $[\bar{1}10]$. This is not, however, established, for in some dislocation reactions similar nodes may, in fact, be joined by a short piece of dislocation, which may remain invisible in the micrograph.

correctness of this picture. Just as in the special case of interaction previously considered (fig. 5), when two screw dislocations were lost, an increase in the dislocation density and stabilization of the dislocation network can result from similar interactions, for example by two edge dislocations which approach one another on intersecting slip planes. The two dislocations, which may be imagined as being straight at first, begin to expand into arcs as the stresses on them are increased and, before reaching positions at which they would become unstable and increase their lengths by operating as Frank-Read sources, they interact to give a Cottrell-Lomer dislocation§ over a small length where they first come into contact. They may, therefore, become stabilized in the expanded state by the immobile Cottrell-Lomer lock, with a resulting increase of total length of dislocation line in the area of their interaction. It should be noted that the operation of Frank-Read sources is not essential in this process, and probably does not occur at all frequently in deformed polycrystalline metals.

The influence of the network on the critical resolved shear stress at which a crystal will begin to flow is reflected in the expression[14]

$$\tau \simeq G\mathbf{b}/\lambda \quad \ldots\ldots\ldots\ldots\ldots\ldots(9)$$

where G is the shear modulus, **b** the Burgers vector of glide dislocations and λ the statistically most probable spacing between locking points on dislocations contributing to the flow. If the stress is subsequently held constant under isothermal conditions the crystal may be observed to creep; in the light of equation (9) this must mean that in the course of time new dislocations fulfil the criterion of equation (9) and contribute to the flow. Isothermal recovery must therefore result in a displacement of the distribution of lengths of network links to larger values, as we surmised above.||

If we are concerned with the creep of polycrystals under tensile stresses, σ, we can modify equation (9) by using the approximate relation[4]

$$\sigma = c\tau,\ 2 < c < 3 \quad \ldots\ldots\ldots\ldots\ldots\ldots(10)$$

§This and similar dislocation interactions in face-centred cubic metals have been studied by transmission electron microscopy;[12] in metals of other symmetry, *e.g.* body-centred cubic, analogous reactions can take place, and have been observed.[13]

||Strictly, one should consider all the segments complying with the equation $\tau + \tau_i = G\mathbf{b}/\lambda_i$, $i = 1, 2, 3, \ldots$, where τ_i is the magnitude of the internal stress field near the i-th segment, due to the presence of neighbouring dislocations. However, τ_i can be either positive or negative, so that on summing both sides over all i's, and taking means, one can expect all τ_i's to cancel out very nearly, leaving us with equation (9).

which, together with equation (9) yields

$$\partial\sigma/\partial t = (-cG\mathbf{b}/\lambda^2)\ \partial\lambda/\partial t \quad \ldots\ldots\ldots\ldots(11)$$

With equations (2) and (11) the isothermal creep rate under constant stress becomes

$$d\varepsilon/dt = (cG\mathbf{b}/\lambda^2\chi)\ \partial\lambda/\partial t \quad \ldots\ldots\ldots\ldots(12)$$

where $\partial\lambda/\partial t$ represents the velocity with which the characteristic separation of adjacent nodal points on dislocations would increase if the metal were held at constant strain. In order to be able to solve equation (12) this velocity would have to be expressed in terms of the velocity of jog migration (equation (6)) (as this facilitates changes of λ); the dependence of the coefficient of work-hardening χ on λ, t and T would also have to be known. An approximate relation between the two velocities can be derived for specific modes of jog migration, *i.e.* as discussed previously with reference to fig. 6; the position is less satisfactory with regard to the functional form of χ.

Steady-state creep

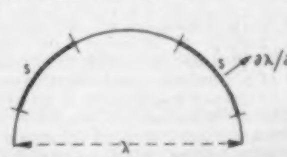
The problem simplifies somewhat if one considers only the steady stage of creep, which is observed at high temperatures after the initial stage of decelerating, transient, deformation. One then has the additional condition

$$d^2\varepsilon/dt^2 = 0 \quad \ldots\ldots\ldots\ldots\ldots\ldots(13)$$

which implies that the rate of work-hardening and the rate of recovery are equal. The 'equilibrium' values of χ and λ, in principle obtainable with this relation from equation (12), are no longer functions of the time t. We shall denote these parameters by χ_e and λ_e, remembering that both are still dependent on T.

Next, in order to find a relation between v_j and $\partial\lambda/\partial t$ we shall use as a rather rough model a single representative dislocation approximately semi-circular in shape, as suggested by the 'Frank-Read' equation (9), in which jogs are held up in the 'mixed' parts, denoted by s in fig. 9.

The passage of one jog over the length s will induce climb by one interatomic spacing and, at the same time, facilitate the expansion of the loop in the slip plane, also by a few **b**. If the mean



9 *'Mixed' screw/edge parts, s, of a nearly semi-circular dislocation of diameter* λ

spacing between jogs on s is l, then the loop will expand due to slip at a velocity

$$\partial\lambda/\partial t = \alpha\, v_j \mathbf{b}/l,\ 1 \leq \alpha \leq l/\mathbf{b} \quad \ldots\ldots\ldots(14)$$

where v_j is as defined by equation (6). This result is to be regarded only as an order of magnitude estimate; further refinement of the model would at present be scarcely worth while.

By combining equation (14) with equations (12) and (4), writing $\dot{\varepsilon}$ for the equilibrium value of the creep rate, one obtains, with $\alpha = 1$:

$$\dot{\varepsilon} = (cG/\chi_e\lambda_e^2)\,\mathbf{b}\,\{(\mathbf{b}/l_e)\,\nu\mathbf{b}\exp\{-[(H-q\sigma)/\boldsymbol{k}T]\} \quad \ldots(15)$$

where

$$q = \mathbf{b}^2 l_e/c$$

the subscripts in each case referring to the equilibrium value as attained in the steady stage of creep. On comparing equation (15) with equation (3) the first bracketed term is seen to relate to the density of active dislocations. Making reasonable assumptions about the magnitudes of the parameters in that term, *e.g.* $c = 2$, $G/\chi_e = 50$, and $\lambda_e = 3 \times 10^{-4}$ cm., one obtains $\rho \simeq 10^9/\text{cm.}^2$ On collecting the pre-exponential terms in equation (15), writing

$$A(T) = \rho\mathbf{b}^3\nu/l_e \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ (16)$$

one finds, with l_e assumed equal to 3×10^{-5} cm., $\nu = 10^{11}$/sec., and $\mathbf{b}^3 = 15 \times 10^{-24}$ cm.^{3} : $A(T) = 50$ sec.$^{-1}$, which is rather less than two times the value of this parameter obtained experimentally in copper at 400°C.[10] If, further, one takes $1/c$ in equation (15) equal to about 0·3, then the value of l_e assumed in obtaining $A(T)$ is also of the same order as obtained by experiment.[15] The magnitude of $A(T)$ increases by a factor of about 2,000 in polycrystalline copper during the transition to a higher activation energy in the range 550—700°C., and this can be accounted for to a large extent on considering that if jogged screw dislocations become rate controlling then $A(T)$ should be $l_e/\mathbf{b}$ times larger than indicated by equation (16).¶

The importance of equation (15), and of similar relations obtainable as indicated in the last paragraph, lies in the fact that it points to the role of the elastic constants and the coarse and fine substructures, represented by λ_e and l_e respectively, in creep. It also provides a useful starting point in considerations bearing on practical problems of the creep resistance of alloys. A major problem which has yet to be solved is the determination of λ_e and l_e in terms of the creep variables. Both appear to increase appreciably with temperature,[10, 15] but the significance of this temperature dependence may have to await further systematic experimental work for its fuller elucidation.

¶If jogs in screw dislocations are rate controlling then one obtains instead of equation (14) the relation $\partial\lambda/\partial t = v_j$; v_j is then associated with an activation energy equal to that of self-diffusion, and not with one approximately equal to vacancy migration as in relation to equation (14). The steep increase in the value of the pre-exponential term in the equation of the steady creep rate in the temperature range in which a pronounced increase occurs in the activation energy has been observed, for example, in copper.[7]

Creep fracture

Extensive work, in particular by high-temperature microscopy,[16] seems to show that the damage eventually leading to creep fracture continues to accumulate throughout the creep process, and does, in fact, appear to be a by-product of the deformation mechanism. This also appears to be borne out by the relation frequently observed to hold for metals and alloys:[7, 10]

$$\dot{\varepsilon} = \varepsilon_0/t_{fr} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots(17)$$

where $\dot{\varepsilon}$ has the same significance as in equation (15), t_{fr} is the time to fracture, and ε_0 a constant having the dimensions of strain.

A process of dislocation 'disintegration' resulting in the formation of small dislocation loops is frequently indicated in transmission electron micrographs of metal foils which have crept. Some of the loops contract and disappear, *e.g.* at 'B' in fig. 7; but occasionally clusters of very small loops are observed which appear to have considerable stability, such as at 'C' in fig. 7. It is possible that the nucleation and growth of holes may be particularly favoured at such spots, especially if they lie close to boundaries or dense dislocation subboundary networks in which a high concentration of vacancies may be expected. Once nucleated, propagation may be favoured by some form of 'Griffith' criterion. Again, however, more specific conclusions will have to await the results of further work.

Acknowledgments

I am indebted to Mr. R. Sinclair for taking the electron micrographs reproduced in this paper.

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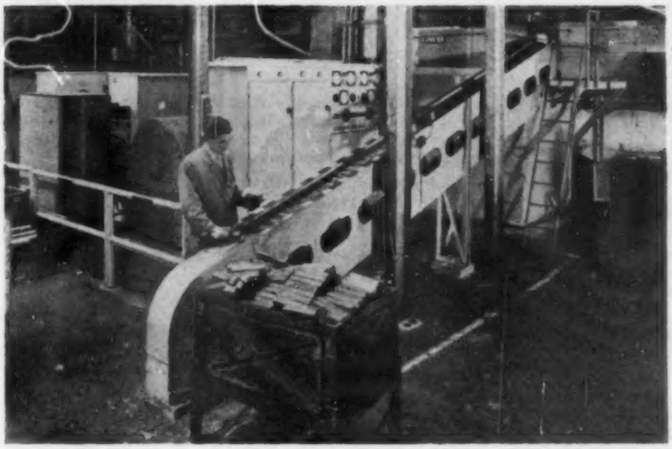


Induction heating of forging billets

To REPLACE three oil-fired furnaces, the Austin Motor Co. Ltd. of Longbridge, near Birmingham, have installed a two-tunnel induction heater for the forging of manganese-molybdenum billets. This new furnace, specially designed by G.W.B. Furnaces Ltd., will do twice the work of the three conventional-type furnaces. Connecting rods for British Motor Corporation engines are finally produced from the billets.

Although the principle of induction heating has been established for some considerable time, it is only in recent years that the many advantages offered by such equipment has been fully realised. The heating achieved by this method is faster than by any other process. It is economic, efficient and the equipment is compact and takes up a relatively small amount of floor space. The plant is clean, and cool working conditions are obtained, an important factor which plays a considerable part in the obtaining of higher production performances by its operators.

All these advantages were considered by Austin engineers before they specified induction heating.

The new furnace, capable of heating up to a temperature of 1,250°C., is wholly engaged in the heating of En 16 steel billets measuring $1\frac{5}{8}$ in. square × $8\frac{5}{8}$ in. long and $1\frac{1}{2}$ in. square × $7\frac{5}{8}$ in. long, each billet making a pair of con rods. Dependent upon which size of billet is being heated, the larger weighs 6·5 lb. and the smaller 4·8 lb. approx.; they can be discharged from each coil at the rate of one every 13 to 14 sec. Pneumatic interlocked pushers operate on a preset time cycle and load one billet into each of the two coils alternately to give a workpiece discharge of one billet/6·5–7 sec. from both coils.

After heating to the required temperature the billets then fall on to an inclined chute which delivers them to the bed of a Wilkins & Mitchell reducing roll. After three rolling operations the billets are hand fed into a Hasenclever forging press where they are flattened, moulded and finished. Finally, they are clipped on a Wilkins & Mitchell clipping press.

The coils, which are the hearts of the furnace, are formed from high-conductivity water-cooled conductors. Each coil is 58 in. long.

Electrically, the billet heater is arranged as two independent units and can be run as a single-tunnel equipment if required. The control panel, and similarly the power equipments, are arranged in the form of two separate sets.

The panel, designed by G.W.B.'s Furnace Division and built by the company's Control Gear Division, houses the mechanical handling control equipment, the alternator regulators and the automatic power factor control equipment for each of the twin coils.

Automatic power factor correction is effected by switched capacitors mounted within the base of the furnace units, which also house the main isolating contactors and water-cooled matching transformers.

The coil units are mounted above this control panel and stand about 8 ft. from the ground. The cropped billets are loaded at floor level on to a flighted conveyor which delivers them to the coil charging position.

Two 300-kW., 2,500-c./s. motor alternator sets are incorporated, each consisting of a 485-h.p. induction motor mounted on a common bedplate with one alternator.

BOOKS

Foundry moulding sands of India

By Mohan, Krishnan, Nijhawan, Gupte and Somayajulu. Council of Scientific and Industrial Research, India. Pp. 180, 16 illustrations, 147 tables. 30*s.* net (15 *Rs.*).

HEAVY INDUSTRIAL development in India is now well advanced, and future projects are planned for the near future. The Indian foundry industry is in the process of being geared to these developments to cope with expansion of existing plant and demands from new industries.

The common factor to the foundry industry is sand and clay suitable for mould and core making and, under the above conditions, greater quantities of known quality materials will be required.

The book takes the form of a monograph, tabulating the necessary data of many suitable sands and is divided, broadly, into two sections. Firstly, a general introduction of four chapters dealing with occurrence, basic characteristics and methods of testing, the latter describing in some detail the individual test procedure (although 10 blows of the standard rammer for making the shatter index test-piece is not routine practice).

The information given here forms the terms of reference for the second section of six chapters, tabulating and reporting specific data of 21 natural moulding sands, 11 crude-silica sands, 10 high-silica sands, one special sand (zircon) and four bonding clays.

It is this second section that justifies publication. Each sand is dealt with separately, giving its geographical occurrence and the name of the company through which it is obtained. Physical and petrological examination, fusion and/or sintering point and moulding characteristics are dealt with exhaustively but concisely and, by way of a conclusion, recommendations are made as to its suitability to make castings of different types and materials.

Macrophotographs at 25 magnifications of each of the sands are reproduced quite clearly and give a good impression of the sand under investigation. Cumulative grading curves likewise help in this way, although interpretation from such a curve is difficult, the more simple block or line graph being preferred by the reviewer. However, the actual size grading percentages for each sieve number is also given, so that relevant information could be replotted to individual preferences.

High-temperature performance is adequately reported by means of tabulated data and photographs of prepared test-pieces fired at different temperatures.

Moulding properties of the silica and special sands pre-suppose the additions of clay, water, etc. A standard addition of 5% Bihar bentonite with variable H_2O content puts these different sands on to a common footing for comparative purposes. The data is clearly tabulated.

Similarly, the chapter dealing with bonding clays uses a washed Rajmahal high-silica sand as a control, with variable clay and water additions.

Both the Bihar bentonite and the Rajmahal sand are dealt with in their respective chapters.

The presentation of such a mass of detail always sets problems to both publisher and printer. It is thought that the individual sections within the chapters could have been made more distinct and the absence of an index makes the search for a single piece of information unnecessarily prolonged. Also the reviewer wonders how familiar even Indian nationals are of the geography and geology of their own sub-continent; a few clear maps would greatly assist in this respect.

Such a volume can have little value outside the field to which it is directed, but to the Indian foundry industry it would appear to be an absolute necessity, especially as it is intended to issue further booklets from time to time on similar lines as new sand and clay deposits are discovered and developed.

T. H. L. BOND

Physical metallurgy

By C. Ernest Birchenall. McGraw-Hill Book Co., New York, 1959. Pp. 323. £3 6*s.* net.

Physical metallurgy

By Bruce Chalmers. John Wiley & Sons, New York, 1959. Pp. 468. £5 net.

THESE TWO BOOKS cover very much the same ground and are directed at much the same audience—university students of metallurgy who are starting the study of physical metallurgy. Both books claim to introduce the new concepts of physical metallurgy that have become of great importance in the last few years, *e.g.* imperfections of crystals, the new concepts of strain-hardening, nucleation and so forth. Prof. Chalmer's book is the longer with 468 pages against 323 of Prof. Birchenall's book. Prof. Chalmer's book is rather more practical since it mentions such matters as continuous casting, shell moulding, machining and such practical matters. It is doubtful whether such mention is worth while since presumably any student starting the study of physical metallurgy in a university is well aware that 'a sand mould is made in two parts, each part being moulded on a pattern which represents the appropriate part of the required shape. ' If such a student by some mischance was unaware of this fact he would be unlikely to consult a text on physical metallurgy to remedy the deficiency.

There are other minor differences. Prof. Birchenall allows 25 pages to describe classical and X-ray crystallography. Prof. Chalmers hardly deals with this aspect of physical metallurgy and, in fact, crystallography as such is not mentioned in his index. Prof. Chalmers covers fatigue in six pages—Prof. Birchenall in about two. Prof. Chalmers gives half a page to hydrogen embrittlement, quoting Zappe's theory of high-pressure bubbles without mentioning that Troiano and others dispute this theory; Prof. Birchenall does not mention hydrogen embrittlement.

Both books can be recommended as succeeding in what they set out to do, namely to give an introduction to the modern concepts of physical metallurgy, but, of course, can only give an introduction. Both are well printed and clearly illustrated. The reputations of both authors are a guarantee of the soundness of the texts. Which is brought by a student is a matter of taste; the reviewer prefers Prof. Birchenall's, but would be hard put to justify his choice.

J. H. RENDALL

Electrical Aids in Industry—Data Sheet **19**

Factory Heating—2

It is impossible in this Data Sheet to describe in detail every type of electric heater on the market, but a representative selection is dealt with below. Each type of building presents its own problem, and the best plan is to seek advice from your Electricity Board, who will be pleased to help.

'OFF PEAK' ELECTRIC HEATING

Because the 'off peak' load makes use of generating and distributing equipment when it would otherwise be idle or underloaded, the Electricity Boards offer cheap 'off peak' tariffs. Three types of 'off peak' heating systems are available, namely:

(a) Hot water storage heating: This consists of a conventional hot water radiator or panel heating system through which hot water from a large storage vessel is circulated. The water in the storage vessel is heated electrically during the 'off peak', low tariff hours and is circulated when required through the radiators or panels.

(b) Block storage heaters: These heaters consist essentially of a number of firebrick blocks which are heated up during the 'off peak' hours by means of suitable electric heating elements. The storage heaters are clad with a layer of suitable heat-insulating material and are housed in a sheet metal casing, the design being such that the stored heat is gradually dissipated throughout the day by means of radiation and convection. These heaters can easily be installed in existing buildings.

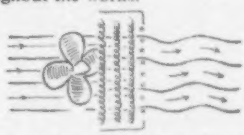
(c) Floor warming: In an 'off peak' floor warming installation, electric heating

Plan view of room *Isometric view of heating cable*

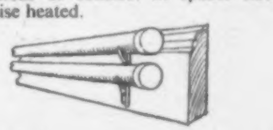
cables or ducts housing withdrawable cables are embedded in the concrete floor of the building. The cables are switched on and the floor is heated up during the 'off peak' hours, and the mass of concrete and screed of the finished floor has sufficient thermal storage capacity to heat the building during the period when current is not available. This method is only applicable to new buildings.

DIRECT ELECTRIC HEATING

(a) Unit heaters: These consist of a bank of electric heating elements fixed in a casing on which is mounted a fan which draws or blows air over the heating elements and discharges it in the required direction. Such units are mounted on the walls or stanchions or hung from the roof members in appropriate positions throughout the works.



(b) Infra-red heaters: These consist of heating elements usually of the sheathed metal or silica tube type mounted in a polished reflector. They operate at temperatures from 700° to 900°C, and give off the greater part of their heat output by radiation. They are mounted overhead and are particularly useful for providing local areas of comfort in spaces not otherwise heated.



(c) Tubular heaters: These take the form of tubes approximately 2" in diameter containing an electric heating element and are available in lengths from 2 to 17 feet. The normal loading is 60 watts per foot run and the surface temperature is from 180° to 200°F. They are usually placed round the walls at skirting level, but also can be used at high level in order to prevent downdraughts.

For further information get in touch with your Electricity Board or write direct to the Electrical Development Association, 2 Savoy Hill, London, W.C.2. Telephone: TEMple Bar 9434.

Excellent reference books on the industrial and commercial uses of electricity and reprints of articles and papers are available.

E.D.A. have available on free loan in the U.K. a series of films on the industrial uses of electricity. Film and Book catalogues and Publications List sent on request.

CUT HERE

NEWS

The I.S.I. in the U.S.

ABOUT 250 MEMBERS of the Iron and Steel Institute and guests have been in the U.S. for the special meeting of the Institute. Most were, of course, from the U.K., but other countries were well represented, including some members from Egypt, India, and Japan. The visitors were welcomed last month at the new United Engineering Centre of the American Institute of Mining, Metallurgical and Petroleum Engineers, New York City, by Mr. J. S. Smart, Jr., president of the AIME.

Mr. Ernest Kirkendall, secretary of AIME, made the announcement that Sir Charles Goodeve had been elected to life membership of the AIME.

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£4½ million expansion scheme by Firth-Vickers

Firth-Vickers Stainless Steels Ltd., of Sheffield, announces a scheme for modernizing and reorganizing their facilities for the production of plates and wide sheets. The scheme will cost in the region of £4,500,000 and take three or four years to complete, having to be carried out in stages to avoid undue interference with production.

The first stage is bascially the replacement of obsolete mills by a new 80-in. wide hot reversing mill with fully mechanized ancillary equipment for the production of plates up to 6 ft. finished width and from $\frac{3}{16}$ in. to 1 in. thickness. The mill will also be capable of rolling sheets up to 6 ft. wide which will subsequently be further reduced in thickness by cold rolling.

Plate production, including the mill, will form a new plant to be erected at Shepcote Lane, Sheffield. Civil engineering work in preparation of the site has already commenced.

Completion of the installation of this plant will be followed by a reorganization of the facilities for cold rolling and final processing of wide sheets at the company's Staybrite Works, extending the handling capacity of the plant from the present 48 in. to 72 in. wide.

This scheme is the latest of a series covering the modernization of the company's equipment for the production of stainless steel in its various forms which have been undertaken since the end of the last war.

Continuous casting plant for Tube Investments

Following inspection of existing continuous casting plants in England and on the Continent, Tube Investments Ltd. have decided to install a Concast continuous casting machine at their Round Oak Steel Works, Brierley Hill, Staffs.

The machine will be the pilot plant for experimental and research work for special products for the whole T.I. group. It is to be built and erected by Distington Engineering Co. Ltd., Workington, a subsidiary of the United Steel Companies Ltd., and associates of Concast A.G., Zürich, in the field of continuous casting.

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COMPANY NEWS

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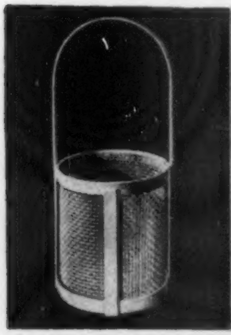
The manufacturing resources of the new company comprise the Electrical Engineering Works at Witton, Birmingham, and the Mechanical Engineering Works at Erith, Kent. The total number of employees exceeds 15,000.

The board of directors of the new company will operate from headquarters in London, negotiations for the acquisition of suitable premises for which are at present in progress.

Hard cash saved in the furnace

WITH Inconel* HEAT-RESISTING ALLOY

INCONEL furnace equipment helps to keep down production costs for the National Cash Register Co. (Manufacturing) Ltd, during heat-treatment of components for their accounting machines, adding machines and cash registers.
Furnace racks of special design in INCONEL alloy are employed during gas-carburising to permit free circulation of the furnace atmosphere around the metal parts to be hardened. Low mass and small heat content of the strong, lightweight INCONEL racks appreciably cuts furnace heat-losses and saves operational costs. The alloy's excellent resistance to oxidation and scaling also increases the working life of racks, thus avoiding frequent replacement and maintenance hold-ups.



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(RIGHT) *Inconel alloy furnace racks in use.*

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Design engineers and others interested are invited to—

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To Henry Wiggin & Co. Ltd.,
Holmer Road, Hereford.

Please send me a free copy of your new publication: 'Wiggin Heat-resisting Alloys'.

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HENRY WIGGIN & COMPANY LIMITED, HOLMER ROAD, HEREFORD

TGA H38C

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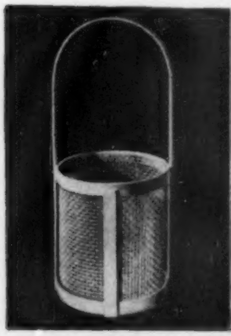

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TGA H38C

PEOPLE

THE COUNCIL of the Iron and Steel Institute has decided to nominate **Mr. M. A. Fiennes,** M.I.MECH.E., group managing director of Davy-Ashmore Ltd., at the Institute's autumn general meeting on November 29 for election at the annual general meeting on May 2, 1962, as president for 1962-63.

Mr. J. B. Simpson, A.C.S.M., M.I.M.M., M.I.MIN.E., mining consultant to Anglo-French Exploration Co. Ltd., has been elected president of the Institution of Mining and Metallurgy for 1962-63.

He was educated at Bedford and, after service in the R.A.F. in 1918-19, was trained at the Camborne School of Mines from 1919 to 1922. Until 1928 he was engaged in tin mining in Cornwall, Northern Nigeria and Bolivia, after which he spent nine years at the Champion Reefs mine on the Kolar Gold Field, South India, and one tour on the Gold Coast at Konongo Gold Mines in 1938-39.

During the war years Mr. Simpson served in the Royal Engineers. At the end of 1942 he succeeded the late Col. L. C. Hill as C.R.E. in command of the 1st Tunnelling Group, R.E., until 1945.

After the war Mr. Simpson joined New Consolidated Gold Fields as assistant engineer in the London office and remained there for ten years, ultimately succeeding Mr. A. R. O. Williams as resident engineer.

Mr. Simpson joined the Institution as a student in 1921, and was elected an associate member in 1927 and a member in 1944. He served on the council from 1951 to 1958 and has held office as a vice-president since 1959. He has served on the management committee of the Benevolent Fund for the last 11 years, as chairman since 1956. He is a governor of the Camborne School of Metalliferous Mining.

He will take office in May, 1962, in succession to Mr. A. R. O. Williams, O.B.E., A.R.S.M., B.SC.

The council of the Industrial Welfare Society announced the appointment of a new director after a special meeting. **Mr. John Garnett** becomes the Society's third director in its 43 years' existence. The vacancy arose earlier this year when Mr. John Marsh accepted an invitation to become director of the British Institute of Management.

Mr. Garnett comes to the Industrial Welfare Society from Imperial Chemical Industries Ltd., where he is currently responsible for the communications section of the Central Labour Department at head office. He joined ICI after demobilization from the RNVR in 1946 and has had wide experience on both the sales and labour sides in several divisions of the company. Immediately prior to his current assignment at head office he was personnel manager, responsible for staff and labour of the Plastics Division.

At the first meeting of the newly-elected council of the Institution of Works Managers after the annual general meeting, held on October 28, the **Rt. Hon. Lord Piercy,** C.B.E., was re-elected president for the ninth successive year.

The outgoing chairman, **Mr. A. M. Hudson-Davies,** O.B.E., M.A., was succeeded in office by **Mr. John Ayres,** M.I.E.E., M.I.P.E., Fellow of the Institution of Works Managers, managing director, Simms Motor Units.

Having been chairman of the Vitreous Enamel Development Council since its formation in 1956, **Mr. S. W. Vickery,** managing director of Ferro Enamels Ltd., has resigned. His place will be taken by **Mr. N. F. Parker,** managing director of Stewart & Gray Ltd. Mr. Vickery will continue to serve as a director of the VEDC.

The chairman of the Griffin & George group of companies announces the appointment of **Dr. Dennis S. Beard,** PH.D., A.INST.P., A.R.C.S., B.SC., to the board of Griffin & George (Sales) Ltd. in the capacity of technical sales director.

Dr. Beard, born in August, 1922, was educated at schools in Essex and London before moving on to Imperial College, London, in 1940 to take a degree in physics. After a period in the Department of Naval Construction as administrative head of the radiographic section he was awarded an I.C.I. research fellowship in the Department of Chemistry at Leeds University (1947-51). During this period he visited the United States to take courses at the Massachusetts Institute of Technology and Columbia University, later returning to take up a post with the National Research Development Corporation (1951-54).

He joined the Griffin & George organization in 1960 from the Norwich City College where he had been a lecturer of physics and mathematics.

Mr. William H. Rigg, B.SC., has been appointed managing director of Firth Cleveland Tools Ltd., a member of the Firth Cleveland Group. He will operate from the company's Tipton Works in Staffordshire.

Mr. Rigg was born in Torquay in 1913 and educated at Malvern College, Worcestershire. He studied for two years at the City and Guilds College, London, from 1931 to 1932, and in the following year received his B.Sc. degree.

He began his career as an engineering student with W. H. Allen, Sons & Co. Ltd. in Bedford, and after gaining experience in production engineering with several firms joined Vono Ltd., of Tipton, as production manager in 1937.

Two years later he was made works manager, then a director in 1941, and acting managing director in 1942 at the age of only 29 years. At that time he also became a director of Duport Foundries Ltd., which with Vono Ltd. subsequently merged in the present Duport Group.

At the end of the Second World War Mr. Rigg decided to leave and set up his own specialist team of engineering consultants. Over a period of six years he successfully carried out contracts for industrial concerns and Government departments, including the design and manufacture of special foundry equipment, electronic instruments and high-speed cameras; and the design of furnaces and plant for steel tube construction.

In 1951 he accepted an offer to join Vono Industrial Products Ltd. (now known as Duport Ltd.) as chief engineer to advise and co-ordinate new projects and technical research in the Group. Two years later he was invited to take over direction of his former position with Vono Ltd., which by then had considerably expanded. He was appointed director and assistant managing director, in which position he gained extensive experience in the marketing, sales and distribution of consumer products on a national scale.

In the transitional period while Revo Electric Co. Ltd. was being acquired by the Duport Group, Mr. Rigg was given the task of reorganizing the structure of the company as works director, and in 1957 was appointed managing director.

Mr. R. E. Ansell has assumed the title of sales director

and Mr. C. Bowles has been appointed to the position of sales manager, responsible to the sales director, of Henry Wiggin & Co. Ltd.

Mr. Rene D. Wasserman, chairman of the Castolin-Eutectic companies, celebrated his 50th birthday on November 4 with a trip that will take him to five continents inspecting the business he has built up during his lifetime.

At the age of 29 Rene D. Wasserman sailed for New York from a family business at Pully, Switzerland, to exploit the specialized welding processes that were making the name of Castolin prominent throughout Europe. There he founded an affiliated company, Eutectic, which under his leadership expanded across the continent enlisting the principal industrial companies of North and South America among his clients.

The British Iron and Steel Federation announces the following appointments:

Mr. E. W. Senior, C.M.G., to be director of the Federation.

Mr. J. B. Cowper to be managing director of British Iron and Steel Corporation Ltd.—the industry's main central trading organization—in addition to his present post of financial director of the Federation.

Mr. J. Driscoll to be assistant director (economics).

Mr. L. J. Gollop to be assistant director (statistics).

Mr. B. S. Keeling to be assistant director (training).

Mr. A. H. Mortimer to be assistant director (commercial).

Mr. K. Donohue to be secretary of the Federation.

B.S.A. Tools Ltd. of Birmingham and Metachemical Processes Ltd. of Crawley announce that they have formed—on an equal partnership basis—a new company, Metachemical Machines Ltd.

While the scientific work of the new company will be carried out at its Crawley headquarters, production will be undertaken initially by the Kemworthy Jig & Press Tool Co. Ltd. of Morden, which Metachemical Machines Ltd. has acquired as a wholly-owned subsidiary.

Classified Advertisements

FIFTEEN WORDS 7s. 6d. (minimum charge) and 4d. per word thereafter. Box number 2s. 6d. including postage of replies. Situations Wanted 2d. per word.

Replies addressed to Box Numbers are to be sent, clearly marked, to Metal Treatment and Drop Forging, John Adam House, John Adam Street, London, W.C.2.

MACHINERY FOR SALE

ETHER COMBINED TEMPERATURE INDICATORS AND REGULATORS, zero/1,000°C. Six instruments.
Ether combined indicator and chart recorder, zero/1,000°C. Two instruments.
Two G.E.C. photo-cell equipments comprising projector, type LLH, control relay, type A, and photo equipment, type MD.
Two process timers by Burrell, Sheffield, zero/30 sec.
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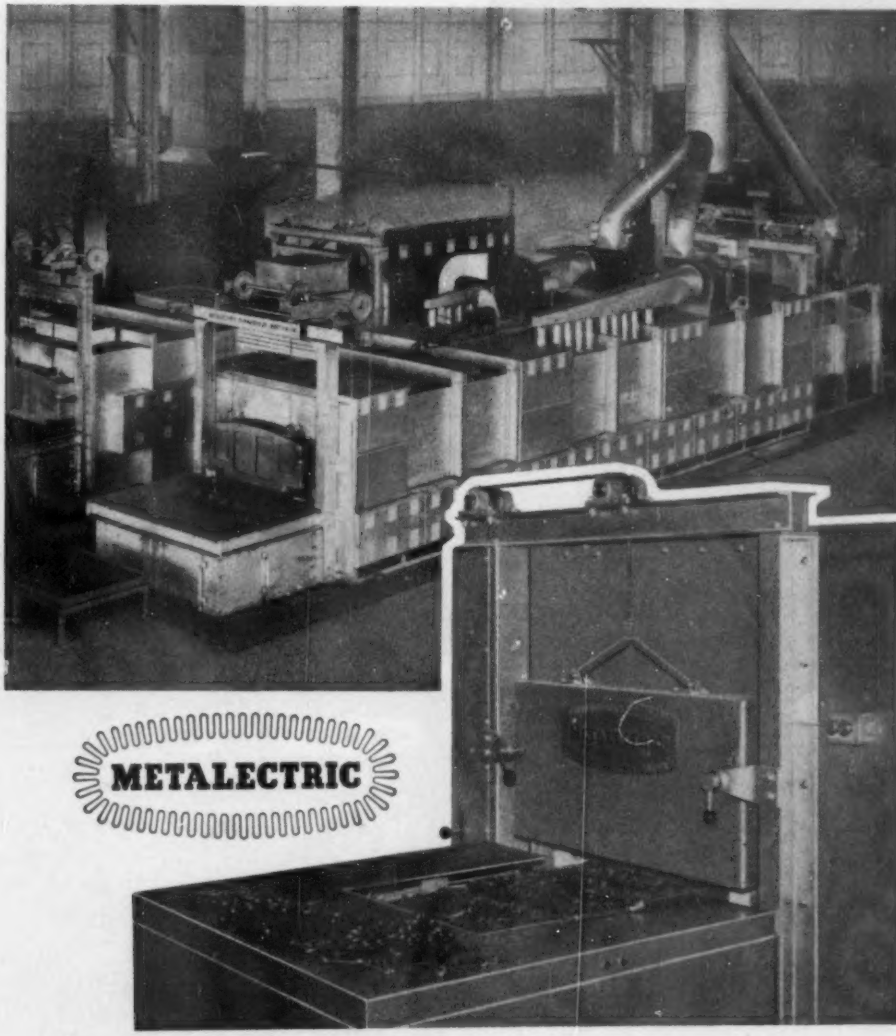

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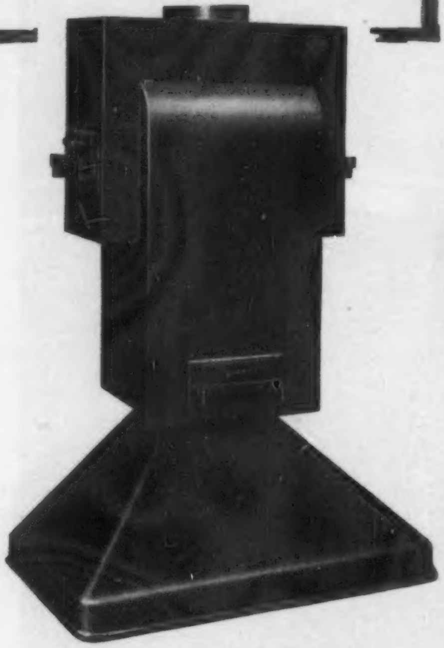

This photograph shows an Alldays & Onions UF/I Dust Collector, primarily for use with hearths and designed to meet the requirements of the Clean Air Act.

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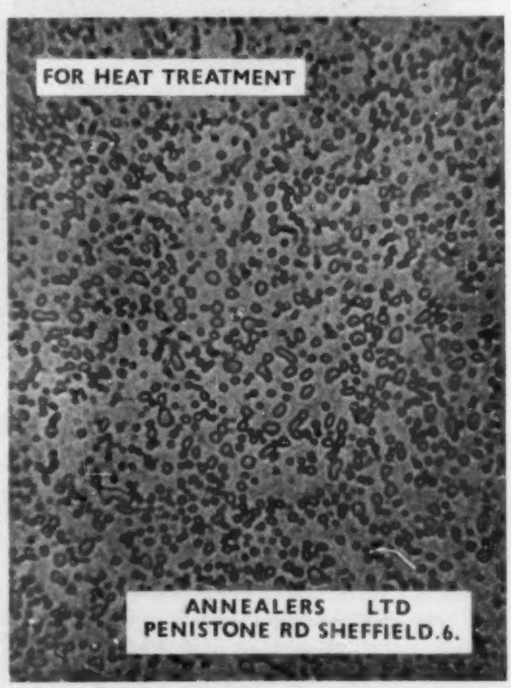

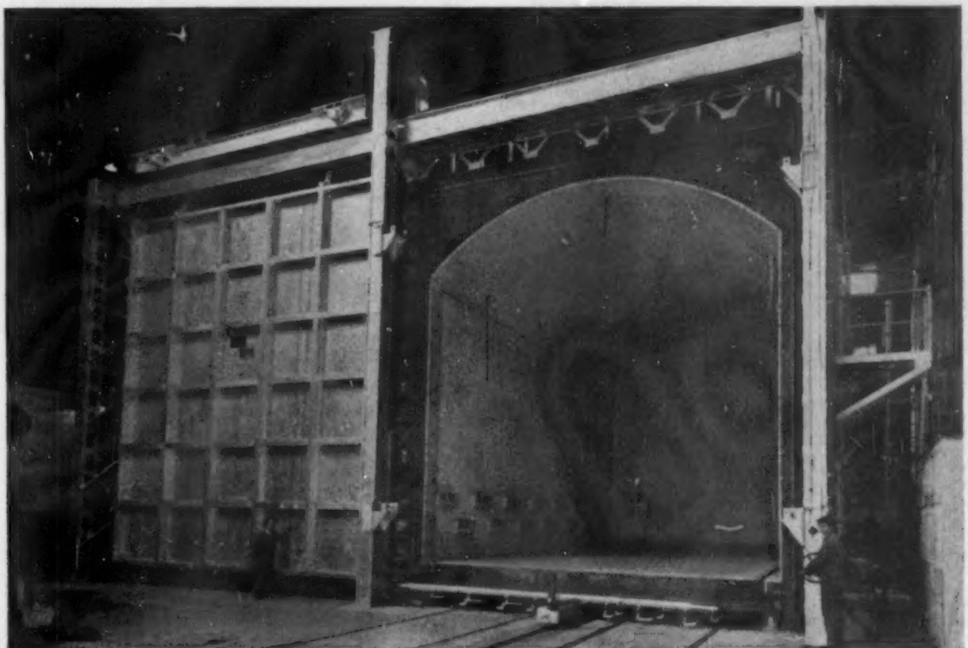

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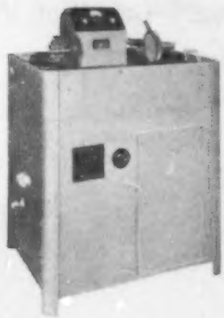

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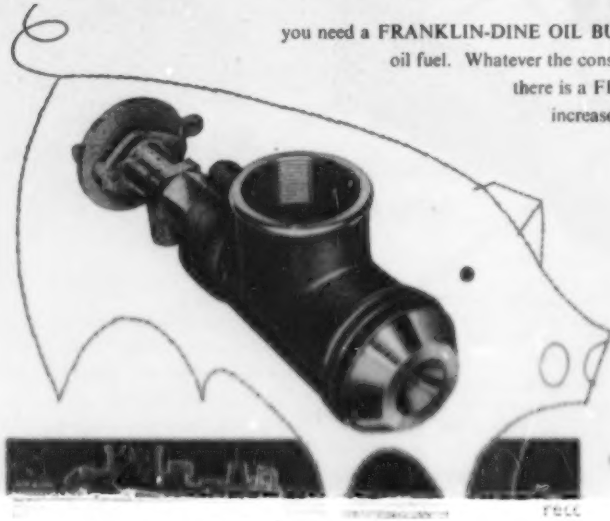

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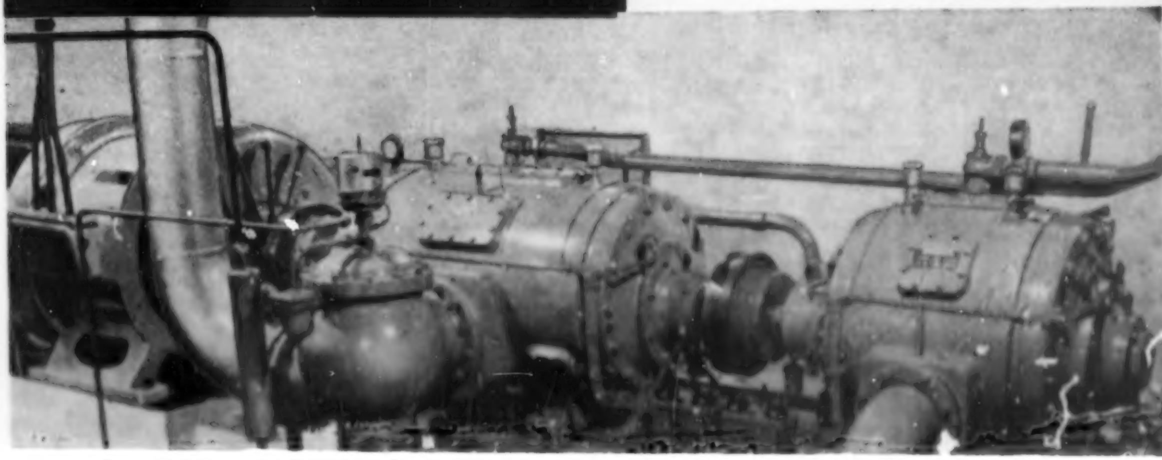

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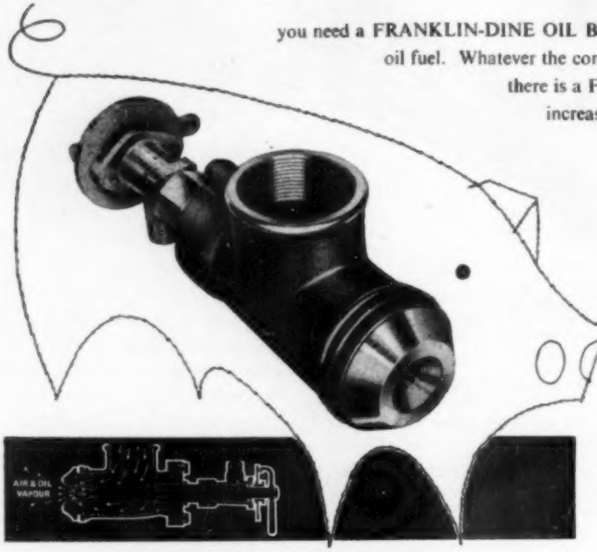

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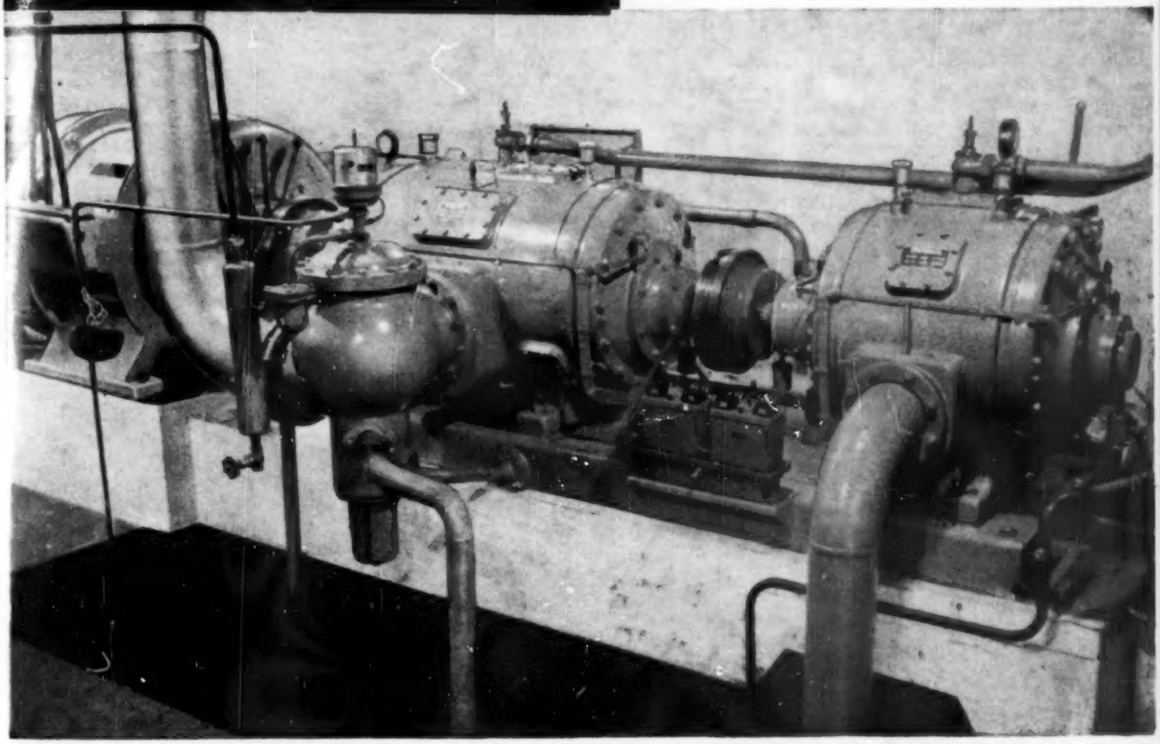

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